TRANSACTIONS

of the American Society for Metals

formerly

American Society for Steel Treating

CONTENTS

ASM

Vol. XXII No. 6
June, 1934

The TRANSACTIONS *of the*

AMERICAN SOCIETY FOR METALS

formerly

AMERICAN SOCIETY FOR STEEL TREATING

Published monthly and Copyrighted, 1934, by the AMERICAN SOCIETY FOR METALS
7016 Euclid Avenue, Cleveland, Ohio

SUBSCRIPTIONS: (members) $2.50 per year
(non-members) $5.00 per year, $1.00 per copy
Foreign (non-members) $6.50 per year, $1.25 per copy

Entered as second class matter, November 9, 1931, at the Post Office at Cleveland, Ohio, under the Act of March 3, 1879.

RAY T. BAYLESS, *Editor*

Vol. XXII　　　June, 1934　　　No. 6

The object of the Society shall be to promote the arts and sciences connected with either the manufacture or treatment of metals, or both.

COMPARISON OF SINGLE-STEP LONG-TIME CREEP RESULTS WITH HATFIELD'S TIME-YIELD STRESS

By A. E. White and C. L. Clark

Abstract

This paper presents data to show whether or not one of the short-time methods advocated by certain metallurgists for determining creep characteristics of metal at elevated temperatures yields results comparable to those obtained from a carefully conducted long-time creep test. Results are included for three steels at 850 degrees Fahr. and eleven at 1000 degrees Fahr.

It is concluded that while the time-yield method does not yield results which are in exact agreement with those from the long-time test, it does offer possibilities as a qualitative test for classifying a series of steels of a given type at any given temperatures.

Introduction

ENGINEERS engaged in the selection and use of metals at elevated temperatures have for a number of years sought a short-time test. Many suggestions have been offered but to date none have proved satisfactory. Dr. W. H. Hatfield, in the Campbell Memorial Lecture in 1928, recommended as a solution his "time-yield" stress. He recommended this again before the British Iron and Steel Institute in September, 1930, and has suggested its use in other papers and addresses. According to his statement[1] "The method . . . consists in discovering, by static loading, the stress within which, at the temperature, stability of dimensions is attained within a period of 24 hours for a further period of 48 hours with an extension not exceeding the elastic deformation by 0.5 per cent on the gage length, with limits as regards measurement for permanence of dimensions to 0.01 per cent of the gage length." In his paper before the Iron and

[1]The Application of Science to the Steel Industry, Campbell Memorial Lecture, American Society for Steel Treating, p. 109-110.

A paper presented before the Fifteenth Annual Convention of the society held in Detroit, October 2 to 6, 1933. The authors are members of the society and respectively Director of the Department of Engineering Research, University of Michigan and Research Engineer, of the same department, University of Michigan, Ann Arbor, Michigan. Manuscript received July 1, 1933.

Steel Institute for September, 1930, he changed the last clause in this definition to read as follows: ". . . and with limits as regards measurement for permanence of dimensions of the order of one millionth of an inch per inch per hour."

From correspondence with Dr. Hatfield, we conclude that his "time-yield" may be determined as the stress which will produce an extension of 48 millionths of an inch per inch of test section between the 24th and the 72nd hour of load application.

For the purpose of throwing further light on the possibilities of this method, a number of steels of different composition and heat treatment were subjected to both the standard creep test and tests to determine the "time-yield." It was hoped that a definite relationship between the values obtained by the two methods would be secured. Beyond question, there is a trend toward such a relationship, but the variations in the two methods are sufficiently great to force engineers seeking relationships with but small percentage variations to conclude that, at least under the conditions of this comparative test, the desired relationships do not exist.

As one studies the proposal of Dr. Hatfield, it is possible that he did not at any time intend to convey an impression of the existence of an exact relationship. He viewed the "time-yield" method as qualitative, but sufficiently quantitative so that by using two-thirds of the value obtained stresses would be secured which could safely be used in engineering design. The outstanding merit to his proposal is that it saves considerable time over the creep methods now in use. It is, without doubt, a most excellent qualitative method. In fact, many of our present-day creep tests can only be so construed, because the values obtained are so dependent on factors outside of chemical composition. For instance, for some steels the creep rate in the spheroidized state has been found to be nearly double the rate when these same steels were normalized followed by a short-time draw.

This explanation accounts in large measure, if not entirely, for the apparent discrepancy in creep values and in Hatfield's time-yield values between the 0.50 per cent molybdenum steel and the National Tube C steel. It also accounts for the apparent discrepancy for these same types of values between the Pittsburgh open steel and the National Tube B steel (which was a killed steel). That is, this latter steel is a killed steel and, for the temperature at which the test was made, it should be superior to an open steel of the

same composition. It did not show this expected superiority. The reason, we feel, is due to the difference in heat treatment preceding the creep test.

With creep rates so subject to variation and with relatively so little work done on the Hatfield "time-yield" method, it is not surprising that a consistent, uniform ratio between the creep stress and the Hatfield "time-yield" value has not as yet been found.

Procedure

All of the steels necessary for this investigation were secured from commercial sources and the majority of them were furnished in the form of ¾ or 1-inch round rods. Their chemical composition and the heat treatment to which they were subjected before testing are given in Table I.

Five of the steels are of the plain carbon type, two of the 0.50 per cent molybdenum type, and four of the 4 to 6 per cent chromium, 4 to 6 per cent chromium plus tungsten, or 4 to 6 per cent chromium plus molybdenum type. Both open and killed plain carbon steels were considered.

Table I
Chemical Composition and Heat Treatment of Steels Used for Comparison of Results from Long-Time Creep Test and Hatfield's Time-Yield Value

		Chemical Composition, Per Cent					
Designation	Heat Treatment	C	Mn	Si	Cr	W	Mo
Grade A	As Received	0.18	0.49	0.01	....	....	....
Grade B	As Received	0.41	0.92	0.21	....	....	....
Nat. Tube B (Open)	N. D.[1]	0.15	....	0.01	....	....	....
Nat. Tube A (Killed)	N. D.[1]	0.15	....	0.20	....	....	....
Pitt. Open	Unknown	0.15	....	0.01	....	....	....
D-1	N. D.[2]	0.15	0.26	0.43	1.24	1.13	....
Nat. Tube D	N. D.[1]	0.15	1.95	....	....	....	0.55
0.50 Mo.	N. D.[2]	0.16	0.47	0.23	....	....	0.42
Nat. Tube C	N. D.[1]	0.15	....	....	....	....	0.50
Nat. Tube E	Ann.[3]	0.15	....	....	4–6	....	....
Nat. Tube F	Ann.[3]	0.15	....	....	4–6	....	0.50
4–6 Cr + W	Ann.[4]	0.08	0.42	0.20	4.88	1.10	....
4–6 Cr + W	Q. D.[5]	0.25	0.38	0.23	5.50	0.80	....

[1]Normalized and drawn at 1200 degrees Fahr. for 168 hours.
[2]Normalized and drawn at 1200 degrees Fahr. for 1 hour.
[3]Annealed and drawn at 1200 degrees Fahr. for 168 hours.
[4]Annealed.
[5]Quenched and drawn at 1200 degrees Fahr. for 1 hour.

All of the long-time creep tests were conducted according to the single-step method of loading. That is, a separate specimen was used for each stress. In the majority of cases, four stresses were used for each steel, and each of these stresses was maintained constant for a period of approximately 500 hours.

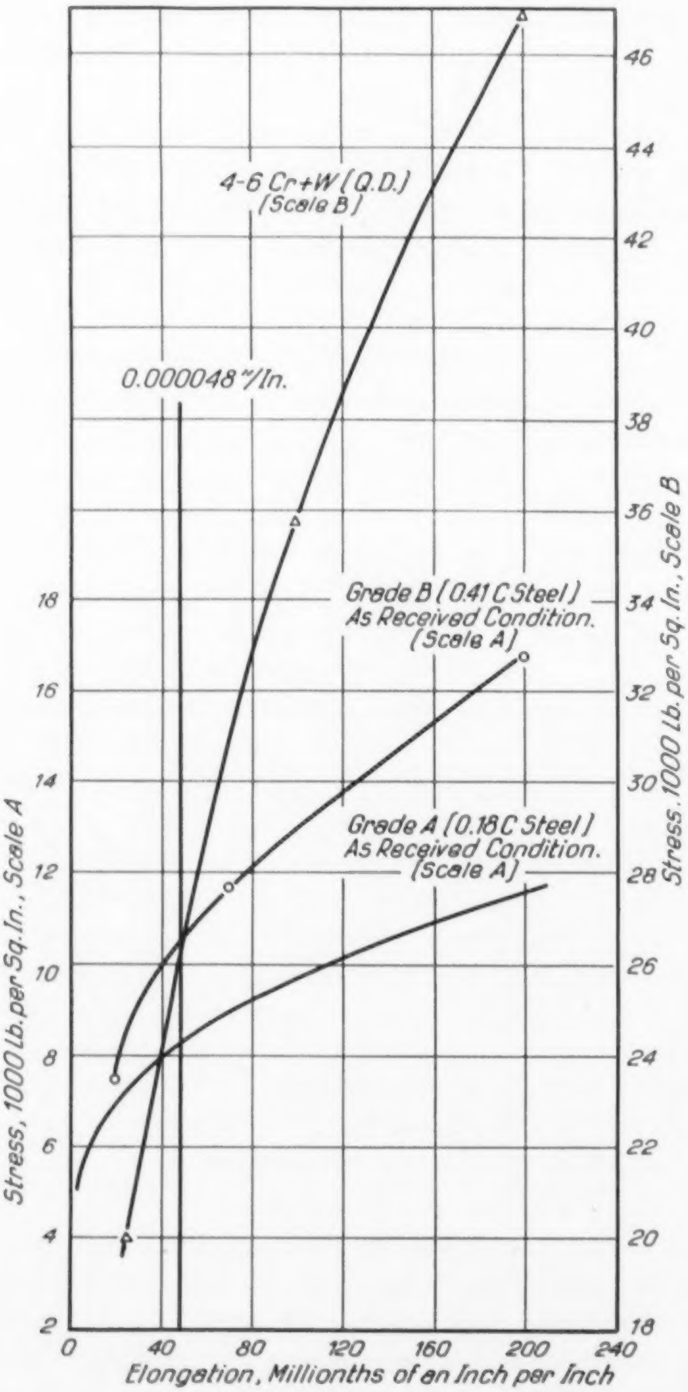


Fig. 1—Deformation During 24th to 72nd Hour-Stress Curves for Steels at 850 Degrees Fahr.

The value of the time-yield stress was computed from the time-elongation curves obtained during the long-time creep tests and was taken as that stress which, during the 48-hour period immediately following the initial 24-hour period, would produce a deformation of 48 millionths (0.000048) of an inch per inch. This value was computed in each case by plotting the deformation occurring during this period against the corresponding stress, and determining the stress at which the curve thus obtained crossed the line designating a deformation of 48 millionths of an inch.

The usual long-time creep results, expressed in terms of the stress required for a rate of creep of 1 and 10 per cent per 100,000

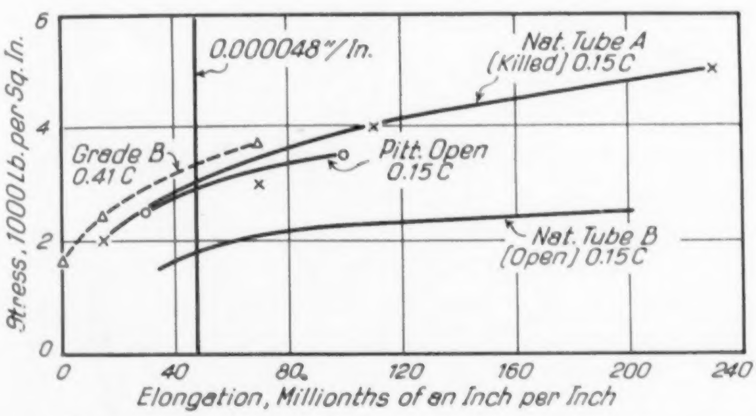


Fig. 2—Deformation During 24th to 72nd Hour-Stress Curves for Plain Carbon Steels at 1000 Degrees Fahr.

hours (0.01 and 0.10 per cent per 1000 hours), were compared with the calculated value of the time-yield and also with two-thirds of this value. This particular fraction of the time-yield was chosen as Dr. Hatfield recommends that the working stress be taken as two-thirds of the time-yield value. In many of our results, however, it was found that this particular ratio did not best suit the findings. Especially was this true for the results obtained at 850 degrees Fahr. In this case, the best average ratio was computed.

Results

The elongation obtained during the 24th to 72nd hour period of the long-time creep tests under the designated stress and the calculated value of Dr. Hatfield's time-yield stress are given in Tables II and III, while Figs. 1 to 3, inclusive, give the graphical presentation of the elongation values and indicate the method employed for

Table II
Elongation for Selected Steels at the End of 24 and 72 Hours of Steady Load Application at 850 Degrees Fahr.

Designation	Type Composition	Stress Lb./Sq.In.	Elongation in Millionths of Inches Per Inch End of 24 Hrs.	End of 72 Hrs.	24-72 Hr. Period	Hatfield's Time-Yield Lb./Sq.In.
Grade A	0.18 C, 0.49 Mn, 0.01 Si	5,000	25	28	3	......
		7,500	220	240	20	......
		7,500	255	300	45	......
		9,225	748	828	80	......
		11,700	1,487	1,697	210	8,300
Grade B	0.41 C, 0.92 Mn, 0.21 Si	7,500	424	444	20	......
		11,700	583	653	70	......
		16,795	1,376	1,576	200	10,450
4–6 Cr + W	0.25 C, 0.38 Mn, 0.23 Si	20,000	885	860	25	......
	5.50 Cr, 0.80 W	35,840	1,900	2,000	100	......
	Quenched and drawn	46,880	2,380	2,580	200	26,250

determining the time-yield. Fig. 1 gives the values for three steels at 850 degrees Fahr. Fig. 2 gives the values on four plain carbon steels at 1000 degrees Fahr. and Fig. 3 gives the values for a number of different alloy steels at 1000 degrees Fahr.

Tables IV, V and VI give the stresses required for creep rates of one and ten per cent per 100,000 hours as obtained from the long-time creep tests, together with the values for Hatfield's time-yield, as well as two-thirds of this stress. The same results are shown graphically in Figs. 4, 5 and 6.

Table IV and Fig. 4 show that at 850 degrees Fahr. the aver-

Table III
Elongation for Selected Steels at the End of 24 and 72 Hours of Steady Load Application at 1000 Degrees Fahr.

Designation	Type Composition	Stress Lb./Sq.In.	Elongation in Millionths of Inches Per Inch End of 24 Hrs.	End of 72 Hrs.	24-72 Hr. Period	Hatfield's Time-Yield Lb./Sq.In.
Nat. Tube B	0.15 C (Open)	1,500	75	110	35	.....
		2,000	100	160	60	.....
		2,500	460	660	200	.....
		3,000	720	1,160	440	1,800
Nat. Tube A	0.15 C (Killed)	2,000	110	125	15	.....
		3,000	240	310	70	.....
		4,000	370	470	100	.....
		4,000	280	390	110	.....
		5,000	450	680	230	3,000
Pitt. Open	0.15 C (Open)	2,500	150	180	30	.....
		3,500	240	340	100	2,950
Grade B	0.41 C, 0.92 Mn,	1,635	109	109	0	.....
	0.21 Si	2,420	120	134	14	.....
		3,700	176	245	69	3,300
Nat. Tube E	0.15 C, 4-6 Cr	5,000	60	120	60	.....
		6,225	190	330	140	.....
		7,500	360	560	200	.....
		9,225	2,400	4,720	2,320	4,400

Table III (Continued)

Designation	Type Composition	Stress Lb./Sq.In.	Elongation in Millionths of Inches Per Inch: End of 24 Hrs.	End of 72 Hrs.	24-72 Hr. Period	Hatfield's Time-Yield Lb./Sq.In.
Nat. Tube F	0.15 C, 4-6 Cr, 0.50 Mo	5,000	10	35	25	
		7,500	380	480	100	
		9,225	560	720	160	
		11,700	1,840	2,686	846	6,400
Nat. Tube C	0.15 C, 0.50 Mo	5,000	30	60	30	
		7,500	245	330	85	
		9,225	620	840	220	
		11,700	800	1,280	480	6,200
Nat. Tube D	0.15 C, 1.95 Mn, 0.55 Mo	5,000	30	50	20	
		7,500	200	260	60	
		9,225	700	920	220	
		11,700	1,340	1,720	380	7,000
4-6 Cr + W	0.08 C, 0.42 Mn, 0.20 Si 4.88 Cr, 1.1 W	6,000	100	125	25	
		7,500	400	520	120	
		10,160	680	1,020	340	6,500
D-1	0.10 C, 0.26 Mn, 0.43 Si 1.24 Cr, 1.13 W	20,000	1,140	1,180	40	
		23,480	1,340	1,460	120	
		26,150	1,535	1,736	201	20,500
0.50 Mo	0.16 C, 0.47 Mn, 0.23 Si 0.42 Mo	9,225	94	99	5	
		14,250	880	1,010	130	
		16,700	2,080	2,280	200	11,400

Table IV
Comparison of Hatfield's Time-Yield Values With Those From Single-Step Long-Time Creep Tests at 850 Degrees Fahr.

Designation	Type Composition	Stress for Designated Rate of Creep Rate=% per 100,000 Hrs. 1.0	10.0	Hatfield's Time-Yield Lb./Sq. In.	66⅔% Time-Yield Lb./Sq.In.
Grade A	0.18 C, 0.49 Mn, 0.01 Si	9,800	12,900	8,300	5,533
Grade B	0.41 C, 0.92 Mn, 0.21 Si	13,000	18,800	10,450	6,700
4-6 Cr + W	0.25 C, 0.38 Mn, 0.23 Si, 5.50 Cr, 0.80 W	25,500	46,500	26,250	17,500

108 Per Cent of average time yield is average creep stress for 1%/100,000 hours.

Table V
Comparison of Hatfield's Time-Yield Values With Those From Single Step Long-Time Creep Tests at 1000 Degrees Fahr.

Plain Carbon Steels

Designation	Type Composition	Stress for Designated Rate of Creep Rate = % per 100,000 Hrs. 1.0	10.0	Hatfield's Time-Yield Lb./Sq.In.	66⅔% Time-Yield Lb./Sq.In.
Nat. Tube B	0.15 C (Open)	1,400	2,200	1,800	1,200
Nat. Tube A	0.15 C (Killed)	1,800	3,400	3,000	2,000
Pitt. Open	0.15 C (Open)	1,800	3,700	2,950	1,966
Grade B	0.41 C, 0.92 Mn, 0.21 Si	2,300	4,800	3,300	2,200

66 Per Cent of average time-yield is average creep stress for 1%/100,000 hours.

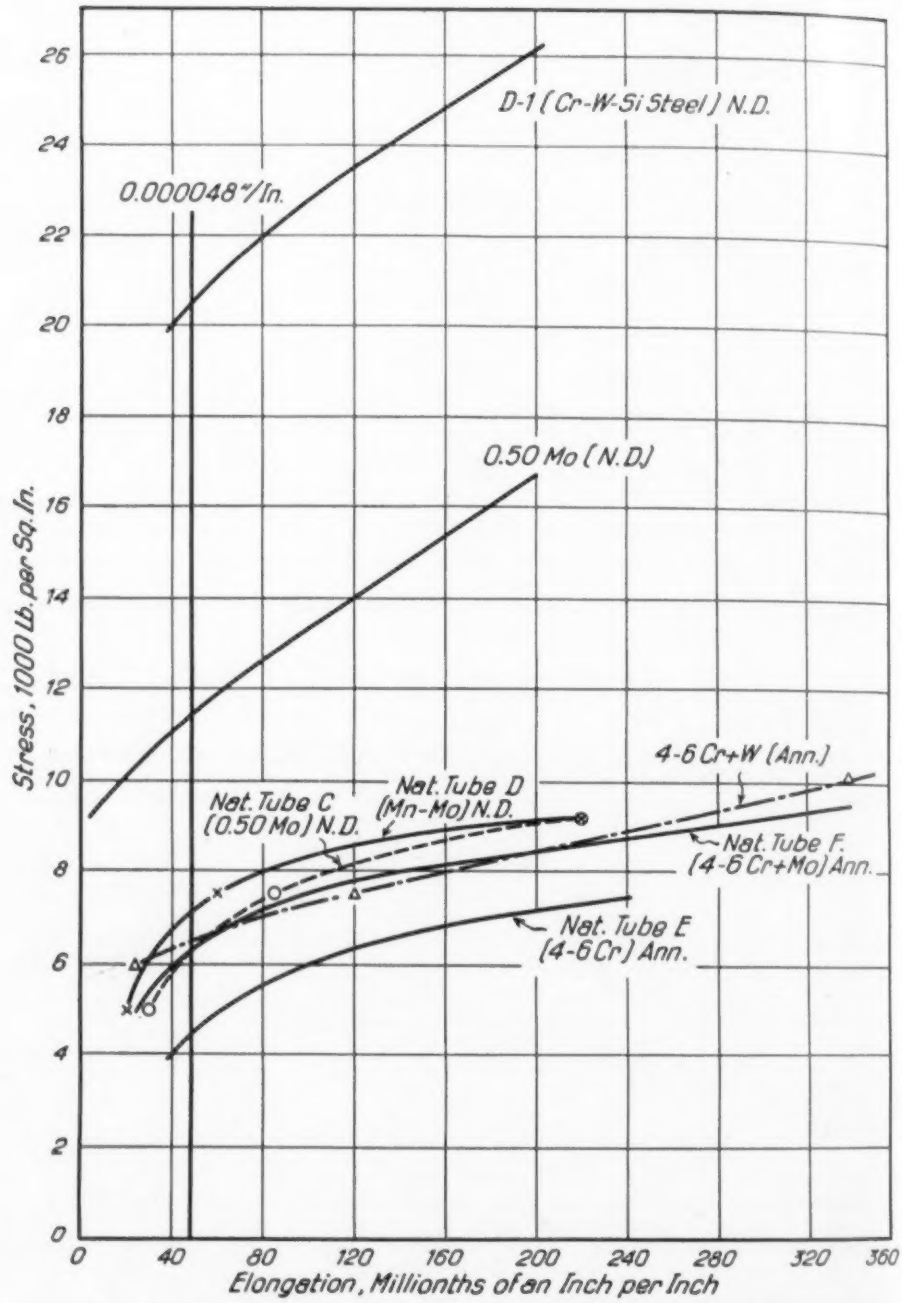


Fig. 3—Deformation During 24th to 72nd Hour-Stress Curves for Alloy Steels at 1000 Degrees Fahr.

age ratio of the stress required to produce a rate of creep of one per cent in 100,000 hours to Hatfield's time-yield is 1.08.

Table V and Fig. 5 give the comparison of Hatfield's time-yield values with those from single-step long-time creep tests at 1000 de-

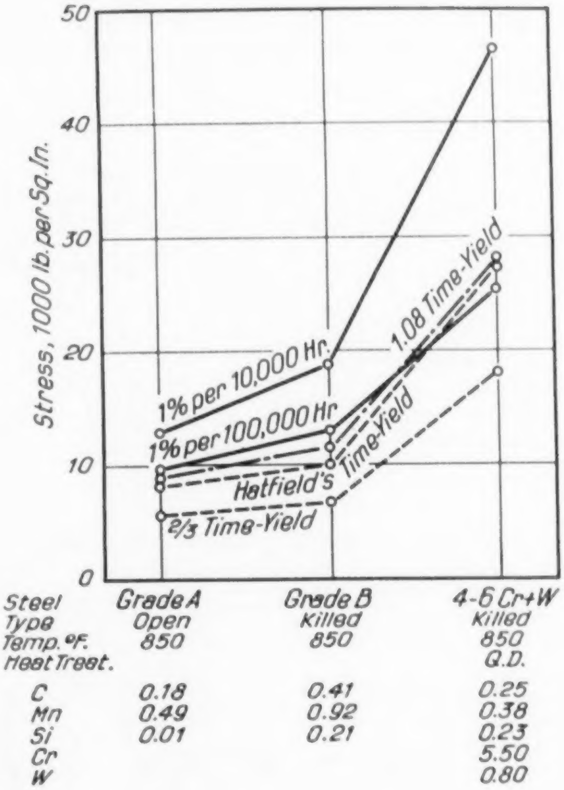


Fig. 4—Comparison of Hatfield's Time-Yield Values with Single-Step Long-Time Creep Results at 850 Degrees Fahr.

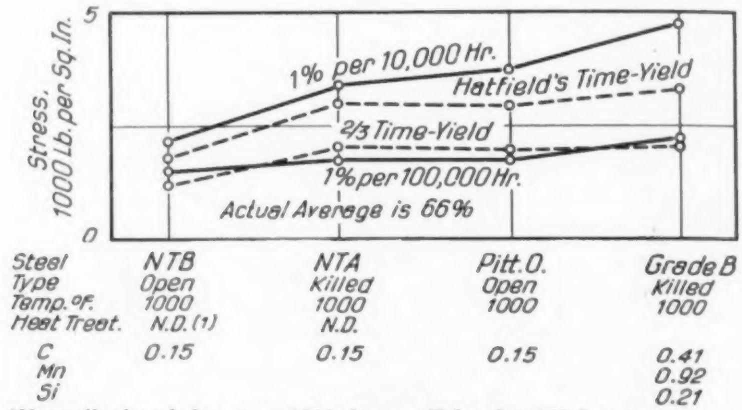
Table VI
Comparison of Hatfield's Time-Yield Values With Those From Single-Step Long-Time Creep Tests at 1000 Degrees Fahr.

Alloy Steels

Designation	Type Composition	Stress for Designated Rate of Creep Rate = % per 100,000 Hrs. 1.0	10.0	Hatfield's Time-Yield Lb./Sq.In.	66⅔% Time-Yield Lb./Sq.In.
Nat. Tube E	0.15 C, 4-6 Cr	3,750	6,600	4,400	2,933
Nat. Tube F	0.15 C, 4-6 Cr, 0.50 Mo	4,200	9,500	6,400	4,267
Nat. Tube C	0.15 C, 0.50 Mo	4,750	8,800	6,200	4,133
Nat. Tube D	0.15 C, 1.95 Mn, 0.55 Mo	5,500	9,200	7,000	4,670
4-6 Cr + W	0.08 C, 0.42 Mn, 0.20 Si, 4.88 Cr, 1.1 W	5,650	8,400	6,500	4,333
D-1	0.10 C, 0.26 Mn, 0.43 Si, 1.24 Cr, 1.13 W	7,300	18,500	20,500	13,667
0.50 Mo	0.16 C, 0.47 Mn, 0.23 Si, 0.42 Mo	9,225	16,000	11,400	7,600

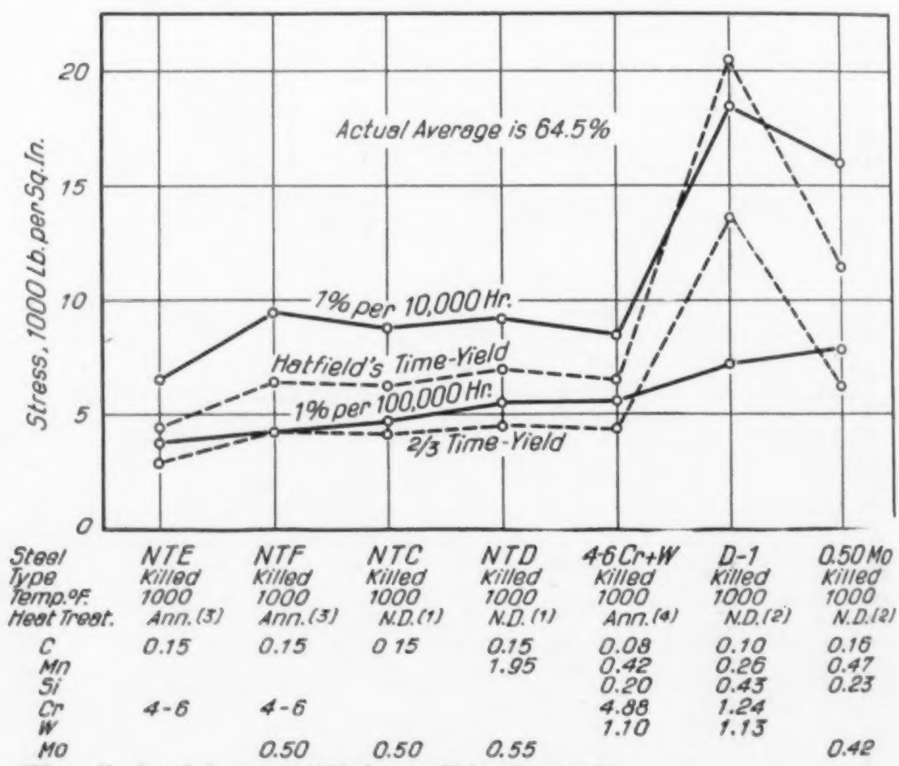
64.5 per cent of average time-yield is average creep stress for 1 per cent for 100,000 hours.

grees Fahr. for plain carbon steels. In this instance, a value of two-thirds of the time-yield closely approximates that obtained from the creep stress.



[1]Normalized and drawn at 1200 degrees Fahr. for 168 hours.

Fig. 5—Comparison of Hatfield's Time-Yield Values with Single-Step Long-Time Creep Results for Plain Carbon Steels at 1000 Degrees Fahr.



[1]Normalized and drawn at 1200 degrees Fahr. for 168 hours.
[2]Normalized and drawn at 1200 degrees Fahr. for 1 hour.
[3]Annealed and drawn at 1200 degrees Fahr. for 168 hours.
[4]Annealed.

Fig. 6—Comparison of Hatfield's Time-Yield Values with Single-Step Long-Time Creep Results for Alloy Steels at 1000 Degrees Fahr.

Table VI and Fig. 6 give the same data for alloy steels. In the case of the alloy steels the approximation is of the same order as that proposed by Hatfield or to be exact, 64.5 per cent. The one alloy steel which is appreciably out of line with this ratio is D-1.

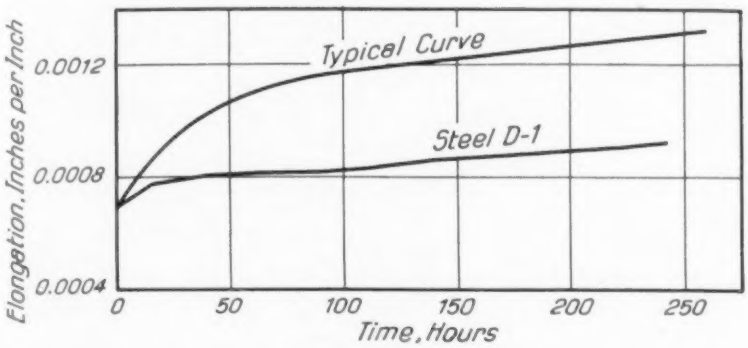
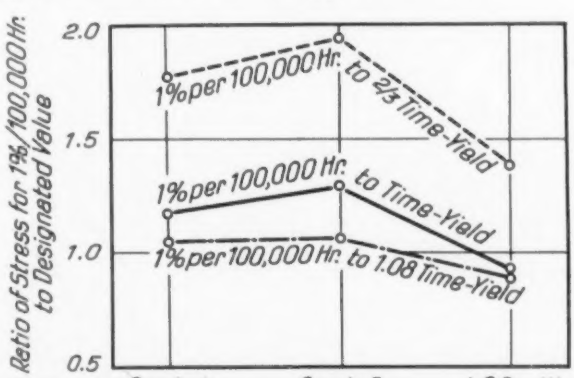


Fig. 6A—Comparison of Time-Elongation Curve for Steel D-1 at 1000 Degrees Fahr. with Curve for Typical Steel.



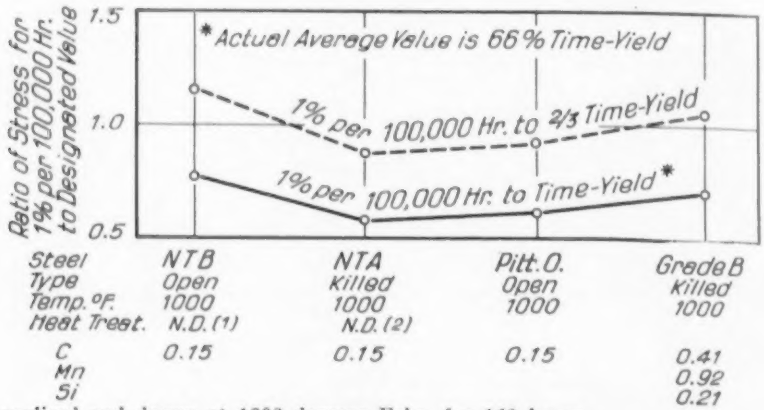
Steel	Grade A	Grade B	4-6 Cr+W
Type	Open	Killed	Killed
Temp. °F.	850	850	850
Heat Treat.			Q.D.
C	0.18	0.41	0.25
Mn	0.49	0.92	0.38
Si	0.01	0.21	0.23
Cr			5.50
W			0.80

Fig. 7—Ratio of Results from Single-Step Long-Time Creep Tests with Hatfield's Time-Yield Values at 850 Degrees Fahr.

This matter has been studied carefully and it has been found that Steel D-1 exhibits a much shorter first stage of creep than the other steels. With all steels except Steel D-1, Hatfield's time-yield is determined entirely in the so-called first stage of creep, while with Steel D-1 it is determined partly in the first stage and partly in the second. This may account for the pronounced difference. The elongation-time characteristics are shown in Fig. 6-A.

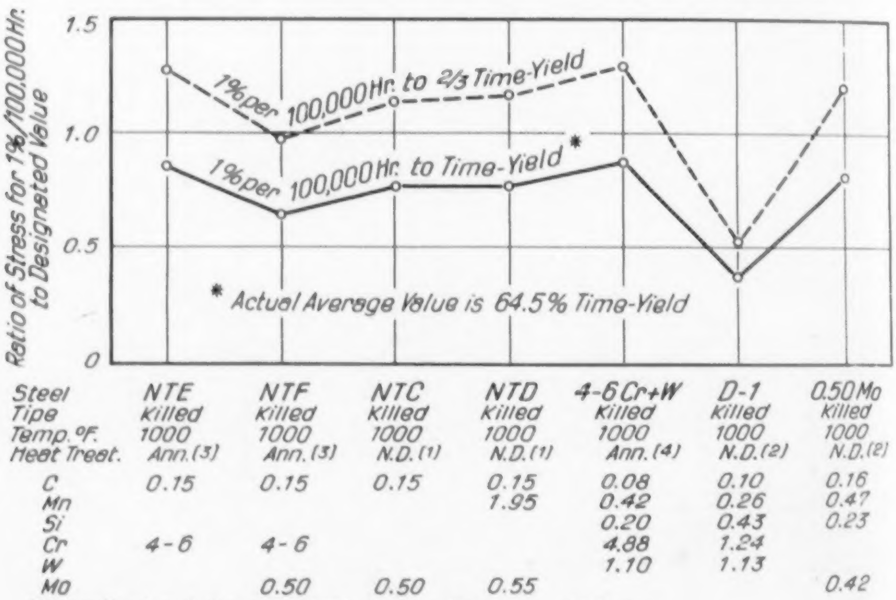
Fig. 7 gives the ratios obtained at 850 degrees Fahr. As stated previously, a value of 1.08 of the time-yield is a better average value for the three steels in question than the ⅔ value proposed by Hatfield.

Fig. 8 gives the ratio for a number of plain carbon steels at



[1]Normalized and drawn at 1200 degrees Fahr, for 168 hours.

Fig. 8—Ratio of Results from Single-Step Long-Time Creep Tests with Hatfield's Time-Yield for Plain Carbon Steels at 1000 Degrees Fahr.



[1]Normalized and drawn at 1200 degrees Fahr, for 168 hours.
[2]Normalized and drawn at 1200 degrees Fahr. for one hour.
[3]Annealed and drawn at 1200 degrees Fahr. for 168 hours.
[4]Annealed.

Fig. 9—Ratio of Results from Single-Step Long-Time Creep Tests with Hatfield's Time-Yield for Alloy Steels at 1000 Degrees Fahr.

1000 degrees Fahr. Hatfield's suggested value of ⅔ is a close approximation, for in this instance the average determined ratio is 66 per cent.

Fig. 9 shows the ratios for a number of alloy steels at 1000 degrees Fahr. Hatfield's suggested average value of ⅔ is a close approximation, as in this instance the average value procured is 64½ per cent. Steel D-1 is considerably out of line but, as stated previously,

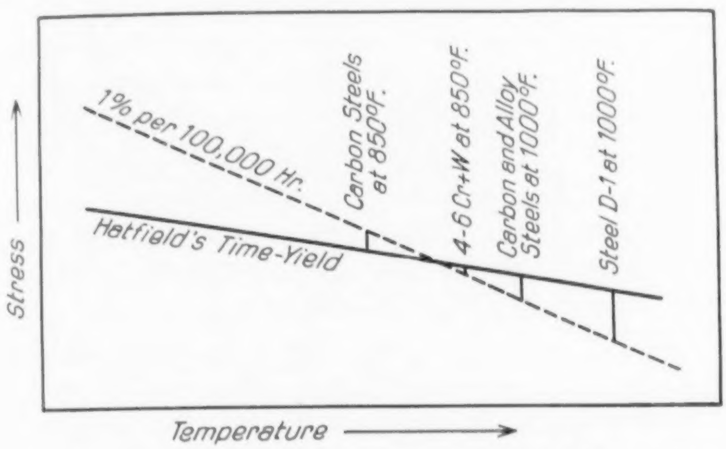


Fig. 10—Diagram of Relationship Between Hatfield's Time-Yield and Stress for Creep Rate of 1 Per Cent per 100,000 Hours at Various Temperatures.

the cause is attributed to the fact that it possesses a short first stage of creep and the time-yield values are computed partly from the first and partly from the second stages, whereas the values used for all of the other alloy steels are those obtained entirely from the first stage of creep.

Fig. 10 is a diagram which shows the relationship between Hatfield's time-yield and the stress for creep of 1 per cent per 100,000 hours at various temperatures. This chart is drawn on the basis of the data contained in this report. It would appear to indicate that a hard and fast ⅔ ratio cannot be used but that the ratio varies at least with temperature, if not with other factors which, in a report of this type, it is not possible to cover.

Conclusions

Results obtained from three long-time creep tests at 850 degrees Fahr. and eleven at 1000 degrees Fahr. conducted by the single-step method of loading were compared with Dr. Hatfield's time-yield stress to determine whether or not a definite relationship existed between the two.

No single relationship was found to exist between the results from these two types of tests. At 850 degrees Fahr. it was found

that 108 per cent of the time-yield stress agreed more closely with the stress required for a rate of creep of 1 per cent per 100,000 hours than the 66⅔ per cent value suggested by Dr. Hatfield. This relationship was not exact, however, for when this particular ratio was used, variations of from 0.90 to 1.07 were obtained.

For the plain carbon steels at 1000 degrees Fahr. it was found that the per cent of the time-yield stress which most nearly agreed with the stress for a creep rate of 1 per cent per 100,000 hours was 66. This agrees closely with Dr. Hatfield's recommended value of 66⅔ per cent. When this ratio is used, however, variations are obtained for the different steels ranging from 1.15 to 0.875.

For the alloy steels at 1000 degrees Fahr., it was found that the per cent of the time-yield stress which most nearly agreed with the stress for a creep rate of 1 per cent per 100,000 hours was 64.5. This also is in close agreement to the recommended value of 66⅔ per cent. However, when this ratio is used the variations for the individual steels were quite pronounced, ranging from 1.30 to 0.55. This lower value is for the chromium-tungsten-silicon steel which failed to show a typical first stage condition. If this steel is omitted, then the range in the ratio values is from 0.975 to 1.30.

As a result of these tests it is concluded that the relationship of creep stress to Hatfield's time-yield varies, depending upon composition and temperature. This is shown in Fig. 10.

It is further concluded that even though the time-yield test does not give values which can be directly converted into actual creep data, it is a method which has possibilities as a relatively rapid qualitative test for the determination of the creep characteristics of ferrous metals, especially in the case of a series of steels of a given type at a given temperature.

Acknowledgments

The authors wish to express their appreciation to The Detroit Edison Co. for the financial assistance given to this study and also The Timken Steel and Tube Co. and The National Tube Co. for sponsoring certain of the creep tests which were used in the composition.

DISCUSSION

Written Discussion: By R. A. Bull, 541 Diversey Parkway, Chicago.

There is great practical significance for several industries, perhaps now particularly for the oil industry, of evaluating proposed methods to determine

the typical tendency of metal to "flow," in service at high temperatures. It is obvious that enormous advantages would accrue from some reliable method that would accurately indicate within a comparatively short time, the behavior of parts in service for several years under high temperature conditions. For all of these reasons many persons will be very appreciative of the work done by the authors, who have incidentally reported valuable comparisons of several important grades of steel.

In wisely emphasizing the number and importance of variable factors which may cause comparisons of testing methods to be misleading, the authors remark that in the case of certain steels the creep rate of the material in a normalized and drawn condition is nearly double the rate when the steel is in a spheroidized condition. Apparently it remains to be determined whether relatively fine granular structure (accompanying a particularly ductile condition attainable in many steels by normalizing and tempering or drawing, and in large measure characteristic of forgings in contrast with castings) is or is not conducive to creep, in all steels of the low alloy grades, in steels that unquestionably are high alloy grades, and in steels which some persons would classify as in a medium group, perhaps touching the imaginary borderlines of high and low alloy varieties. Fairly recent tests indicate that granular structure may have one influence in certain pearlitic steels, and a different effect in some austenitic steels. Unfortunately the problem of determining behavior under continued applications of high temperature seems to become more involved with the comprehensive acquisition of accurate data.

It is gratifying to see skillful efforts made to compare results from an acceptable creep test and the Hatfield time-yield test on grades of steel that now are of particular interest to many consumers. It is possible that the potential value of any relatively short-time test may appeal more strongly to steel foundrymen than to some manufacturers of wrought steel products (including in each case only those producers showing genuine interest in high temperature stability). An important manufacturing element exists, causing casting producers especially to desire the development of a method giving an accurate indication within a comparatively short time for the purpose in mind. Through the acquisition of a great many small melting units during the last 10 years the steel casting industry now is in position to make a vast number of grades of steel without some of the tonnage complications created by the use of relatively large melting furnaces, where the latter are available exclusively. This makes feasible the manufacture of numerous varieties of cast alloy steels, in each of many plants, in a day's work. The progressive foundry equipped to specialize in alloy steels often produces within a single month as many as 30 varieties, and more than that when the plant has one or more very small melting units to supplement equipment of moderately small capacity. Some foundries are prepared to produce commercially and to experiment almost without limit in chemical composition and in heat treatment, dependent only on present market requirements and on reliable test indications that may lead to future demands for the product.

A valuable, interesting contribution that dealt broadly with the subject of plasticity, presumably reflecting the views of a committee of the A.S.M.E., was

presented at the June, 1933, meeting of the society mentioned.[1] This paper outlined five suggested fundamentals as a basis of procedure for making tests to determine behavior of material for seamless steel tubes in the petroleum refining industry. One such fundamental called exclusively for tests of short duration, the cited author giving as the reason for discarding the creep test for the purpose mentioned, the impossibility of approximating hostile service conditions in the laboratory. It was stated in the paper referred to that "There is too much chance of an insidious progressive service deterioration to permit the assumption that creep or any other characteristic of tube material is of a fundamental and enduring nature." Since the views so expressed were applied to purposes to which steels tested by White and Clark are distinctly related, reference to the A.S.M.E. paper does not now seem out of place.

All would agree on the impossibility of introducing into a creep testing unit several deteriorating factors present in oil refinery tubes. Of course these bad actors are also absent when short-time tests are made. It becomes a question of the relative significance of the time element as one of several affecting the working life of the material under test. And it may be taking considerable of a chance to assume even in the case of oil refinery tube material that the time factor is of minor significance compared with other deteriorating factors. These, incidentally, seem to vary considerably, depending on the particular still and on the kind of oil refined.

It will be noted with satisfaction by steel foundrymen and others that the authors, following their comparisons, believe there is considerable utility in the Hatfield time-yield test, when used qualitatively for some steels apparently well adapted for high temperature use. The experiments reported will, it is hoped, lead to others of similar nature for confirmation and amplification; not for the purpose of discounting the creep test, but for practical and supplementary purposes, seriously needed.

There is an important element of time, not only for correctly ascertaining the behavior of material when long exposed to weakening influences, but to the urgent, existing needs of industry. These requirements call loudly for reasonable periods of time to permit the intelligent selection of materials. In a previous publication the writer expressed his belief that "many steel foundrymen think the signs point to the conservation of time as the lesser of two evils, if any relatively short-time method must, ipso facto, carry the hall-mark of doubt, from the ultra scientific standpoint." It might be added here that foundrymen represent but one group of persons seriously interested in the subject, who hold the same conviction, undoubtedly shared by some of the most competent creep test experts; also that the words "ultra scientific" were not used in any derogatory or unappreciative sense.

Written Discussion: By F. H. Norton, Babcock and Wilcox Co., Barberton, Ohio.

This paper brings out clearly the condition which we have for some time believed to exist, namely, that a short-time test will not correlate with long-time creep tests in general. It is true that the particular short-time tests sug-

[1] "Practical Plasticity Problems" (abridged), by G. M. Eaton, *Mechanical Engineering*, Sept., 1933, p. 557.

gested by Dr. Hatfield are in advance of the usual short-time tests where simply the yield point is determined. The former materially increases the time of test and allows some of the strain hardening to be removed. However, in most cases there is a considerable amount of strain hardening up to a time of 72 hours so that we have not by any means reached a condition of stable flow.

It seems probable that in the strain hardening range the rate of flow is influenced to a considerable extent by the previous history of the material aside from its chemical composition. On the other hand, in long-time tests, at least at the higher temperatures, the previous history is more or less erased, and the flow characteristics are determined by the composition of the material.

It may be argued that the long-time creep tests reported in this paper for the various steels, which were of approximately 500 hours duration, were somewhat short to give precise creep values. In some cases the strain hardening range may extend well up to this time. Then again a rather confusing idea of the precision of the results is given when, for example, in Fig. 2 the elongation is stated only to one significant figure, and yet by the time Table IV is reached the results computed from it are given to four significant figures. Fig. 10 is also difficult to make out as apparently the various types of steel are plotted along the temperature axis without regard to their actual testing temperature.

On the whole, this investigation shows that there is some correlation between the short-time test and the creep test, but that there are many discrepancies which would make it unsafe to base any design calculations on tests of this kind. The short-time test might, however, be used to give a rough idea of the value of a new material as the authors suggest.

Written Discussion: By E. S. Dixon, Texas Co., Port Arthur, Texas.

We wish that tests had been made at higher temperatures, say 1200 and 1300 degrees Fahr., because these are the temperatures which metal in the tubes of cracking units attains; and our experience is that failures from creep occur almost entirely at these higher temperatures.

Although any method of evaluating long-time creep by short-time tests is of interest, it will not be sufficient for oil cracking tubes because of the fact that the metal operates at such high temperature for so long a period of time that changes in structure gradually take place so that data obtained on metal in the "as received" condition are not indicative of failures of the same metal when installed in a cracking still. The importance of this point is emphasized by the authors when they speak of the great difference in creep rate of normalized and drawn material, as against that of spheroidized metal.

A criticism of creep work in general is to the effect that data published on any one alloy do not always agree with data previously published on the same alloy in the same heat treated state.

Written Discussion: By Dr. W. H. Hatfield, The Brown-Firth Research Laboratories, Sheffield, England.

Dr. Hatfield wrote that the paper presented by Messrs. White and Clark was of special interest to him, as he thought it was the first description of "Time-Yield" tests ever carried out, except those carried out in the Brown-Firth Research Laboratories, Sheffield, on the basis of which the term "time-yield" was introduced.

He would like to make it clear that in the conception of "time-yield," it was not intended or expected that there should be any definite connection with long-time creep test results. In view of the entirely different basis of evaluation, it is not at all a matter of surprise that there should not be a definite relationship between the time-yield and ultimate creep values—in fact, it is a matter of great interest that the values should follow each other as closely as they do for the different types of steel. If the authors will tabulate the ratios of long-time creep limit values obtained according to the two different rates assumed in their tests for ultimate creep, they will find ratios which have variations of a similar order, although both are based on long-time tests. Here at least one might perhaps have expected a closer relationship. What is the inference? The long-time creep "limit" is still an indeterminate quantity, and in fact can only be defined by specifying, along with the term "creep limit" what rate of creep is implied. The application of the results to practical design cannot yet be considered to have reached a stage where arbitrary rules and personal opinions are unnecessary.

In view of this conclusion, there is plenty of justification for alternative definitions of physical characteristics which can be employed in design, particularly when, as in the case of "time-yield" the physical characteristic is one which can be determined with less labor and in a reasonably short time.

There is another aspect of the question. If one considers the performance of steels under the "ultimate creep" method of testing, at temperatures below those referred to by the authors, and including ordinary atmospheric temperatures, the "ultimate creep" stress finally becomes practically synonymous with "tensile strength." This is obviously not a stress for use in practical design, except by the assumption of a substantial factor of safety—another arbitrary quantity. The slow yielding of steels under stress at high temperatures is, in his opinion, more akin to the ordinary "yield point" than to maximum stress. It was from this point of view that the "time-yield" conception was defined, since it does give a smooth transition connecting high temperature tests with those at lower temperatures.

As to the relative merits of the different methods and criteria of properties of steels at high temperatures, much could be said, but, in his opinion, the important thing is to obtain values which can be a basis of design. A value on which a large factor of safety has still to be allowed does not seem to be the most satisfactory basis. Different engineering applications obviously require different considerations in arriving at safe stresses, and it is clear that in some cases the "amount of initial movement" and time involved before arriving at a specified steady rate of creep, may be as important as the rate of creep itself.

A further point might be mentioned, which is suggested by the exceptional behavior of the sample D.1. in the paper. The data given with regard to the heat treatment of the samples are somewhat incomplete, but it may frequently happen that the temperature at which a piece of steel is tested, or put into service, is one which, independently of the operating stress, is bringing about structural changes in the steel—such, for example, as a slow tempering change. A stable condition for a given working temperature necessitates adequate previous heat treatment, and it seems probable to the present writer that the behavior reported

for sample D.1. may be accounted for by an inadequate previous heat treatment.

In conclusion, he would like to say that in his view the authors will have carried out useful work if they can put the different methods of testing at high temperatures into a proper perspective. He would like to suggest to them that the intermediate range of 300-700 degrees Fahr., as well as the higher temperatures, would be well worth investigation from this point of view.

Written Discussion: By P. E. McKinney, Bethlehem Steel Co., Bethlehem, Pa.

The authors have covered a subject of considerable interest in view of efforts being made by many investigators to find a less time consuming procedure for determining the creep properties of metals.

It appears that the authors are fully justified in their conclusions that no single relationship exists between the results of long-time creep tests and the Hatfield method.

There appears to be some doubt even as to the suitability of this method of test for qualitative evaluation of a series of steels of a given type, since the initial creep in a given type of steel may vary to a considerable degree as the result of different thermal treatments, and such tests would, therefore, not even indicate qualitatively the creep characteristic as developed by a test of one thousand hours or more duration.

Written Discussion: By E. C. Wright, National Tube Co., Ellwood City, Pa.

We are particularly interested in this paper since an important part of the data employed in the discussion originated in connection with tests made for National Tube Co., in the authors' laboratory.

Some of the results presented in the paper are quite striking, particularly the comparative long-time creep values and the calculated Hatfield time-yield value for steels of the same composition originating from different sources.

The specimens marked National Tube Co. A, B, C and D, in Table I, are indicated under the heat treatment as having been normalized and drawn at 1200 degrees Fahr. for 168 hours. Specimens labeled D-1, and 0.50 per cent molybdenum, are also shown to have received a similar treatment, except that the drawing time was one hour instead of 168 hours. It will be noted in Table I that the composition of the 0.50 per cent molybdenum steel and National Tube specimen C is almost identical. By referring to Table VI it will be noted that the creep stress for 1 per cent elongation in 100,000 hours is 4750 pounds for National Tube specimen C, as against a creep stress for the same elongation for the ½ per cent molybdenum steel of the same composition of 9225 pounds, while the corresponding Hatfield time-yield characteristics for these two steels is 6200 pounds and 11,400 pounds, respectively.

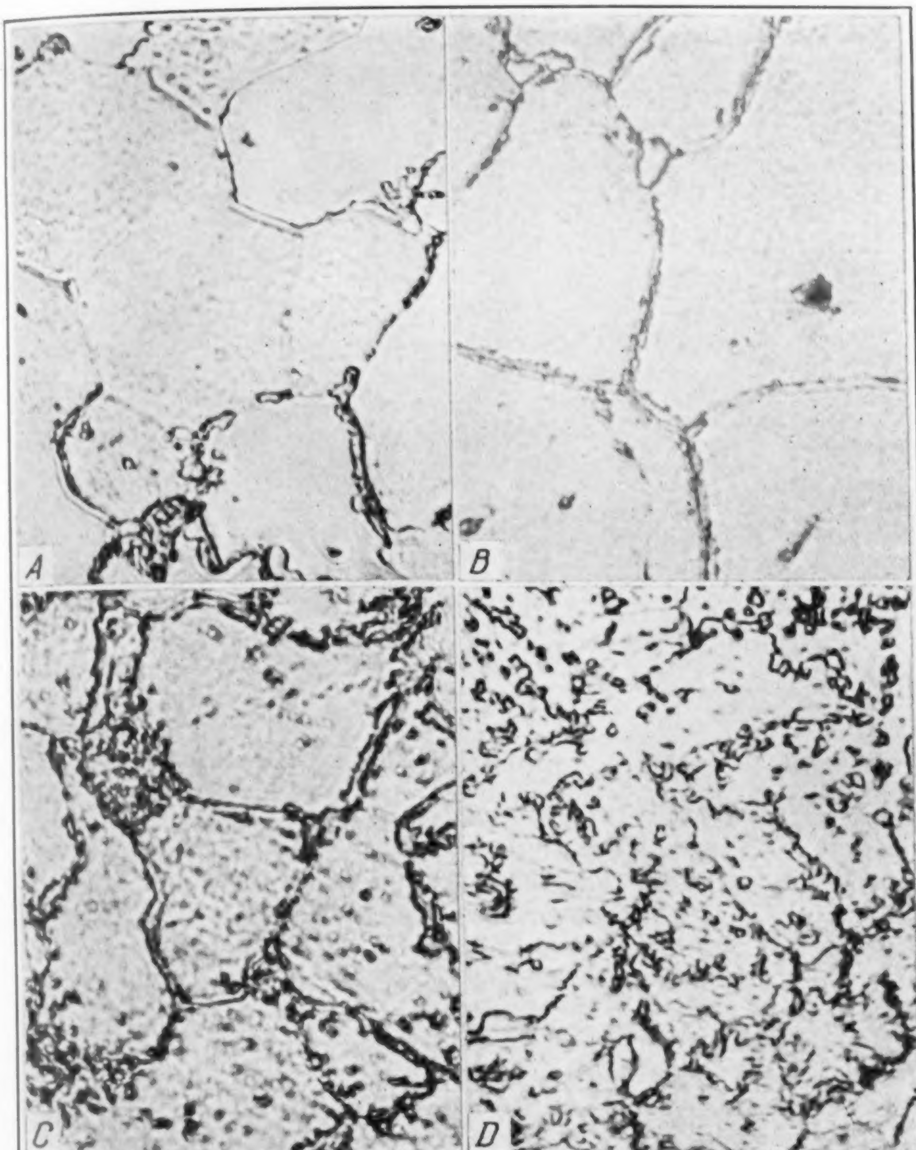
A similar large variation occurs in the comparative properties of National Tube specimen B, a low carbon steel of the rimming type, as compared to the Pittsburgh open steel which we assume to be a rimming steel of the same composition as National Tube specimen as indicated in Table I. By referring to Table V, the creep stress for 1 per cent elongation in 100,000 hours for these two steels is shown as 1400 pounds for National Tube rimming steel and 1800 pounds for the Pittsburgh rimming steel, while the corresponding Hatfield time-

yield value for the two steels is 1800 and 2950 pounds, respectively. In this case the heat treatment of the Pittsburgh rimming steel is reported as unknown. It will be noted, however, that, referring to the same tables, the National Tube specimen A, a killed low carbon steel, has the same creep stress and Hatfield time-yield value at 1000 degrees Fahr. as the Pittsburgh rimming steel; whereas, it is well-known that killed steels of the plain carbon type practically always show higher creep value than open steels of the same composition.

National Tube Co., specimens E and F, also discussed in this paper, were annealed by furnace cooling from 1600 degrees Fahr. and then drawing for one week, 168 hours at 1200 degrees Fahr. The long draw-back temperature on these 4 to 6 per cent chromium steels, which are extremely sluggish in transformation, yielded no apparent alteration in the structure, and, since the specimens were carefully furnace cooled before being subjected to the draw-back, only a slight change in properties would be expected.

The special heat treatment imposed upon National Tube specimens was selected after considerable study of previous creep tests and also several years observation of tubes which had been exposed to high temperature service. It had been repeatedly observed in practice that boiler tubes and refinery tubes which have been in service for several thousand hours or more undergo change in structure, and physical tests made at room temperature on such materials after this service always indicated a considerable softening effect. Similar observations have also been reported by other engineers, notably, Drewry before the A.S.M.E. in 1932, and Bailey before the Institute of Mechanical Engineers of Great Britain in 1932. In view, therefore, of the well established fact that some materials exposed to high temperature service undergo a gradual softening effect under conditions of imposed stress and heat, it was decided to make creep tests on specimens which would approximate this condition as closely as possible, in order to determine what would be the flow characteristics of the steels not when in the new or normalized condition, but after several thousand hours of actual service. It is believed that the heat treatment imposed upon National Tube Co. creep test specimens approximates this condition to a considerable extent.

In order to determine the effect of the long draw-back temperature, tensile tests were made at room temperature before and after the draw heat treatment and prior to creep testing. Table "A", attached, shows the properties obtained, and the softening effect developed by the 168 hours draw at 1200 degrees Fahr. is plainly indicated even in the room temperature tests, being particularly noticeable in steels which have some air hardening characteristics. The temperature of 1200 degrees Fahr. was selected for this draw-back since the creep testing was to be carried out at 1000 degrees Fahr. and Bailey's work indicated that the spheroidization of such steels might be accomplished at this temperature for this period of draw-back. The completeness of the spheroidization obtained for this long drawing treatment will also be indicated by photomicrographs taken of the specimens prior to creep testing. Four photomicrographs are attached showing the structure of National Tube specimens A, B, C and D, and the plates are exhibited at 1000 diameters in order to show the extent of the spheroidization. This structure has been found to be the equilibrium structure



All Specimens Air-cooled from 1650 Degrees Fahr. Drawn 168 Hours at 1200 Degrees Fahr. × 1000.

developed in steels of these compositions in long-time high temperature service.

We believe that this paper is of considerable value in bringing out more strongly than ever the fallacy of making creep tests on materials which do not have a stable metallurgical structure. When results varying almost 100 per cent can be obtained under identical conditions, except as respect to heat treatment, it should certainly indicate that further creep testing on steels, which are undergoing a change in structure even while testing, is unreliable practice for high temperature materials which are to be employed above a temperature of

Table A
Tests at Room Temperature

Type of Steel	Heat Treatment	Yield Point	Tensile Strength	Elongation in 2 inches	Brinell
0.12% Carbon (1)	1650°F.—Air Cool	37,500	53,100	47.0	109
Killed Steel Open Hearth	(1650°F.—Air Cool (1200°F.—168 hrs. draw	35,400	47,200	48.0	102
0.11% Carbon (1)	1650°F.—Air Cool	35,050	50,700	49.0	107
Open Steel Open Hearth	(1650°F.—Air Cool (1200°F.—168 hrs. draw	32,850	45,600	50.0	102
0.16% Carbon (2) 0.50% Molybdenum	1650°F.—Air Cool	42,100	62,300	42.0	122
Killed Steel Open Hearth	1650°F.—Air Cool (1200°F.—168 hrs. draw	37,300	53,800	44.0	114
0.18% Carbon (3) 1.90 Manganese	1650°F.—Air Cool	108,200	129,700	21.0	242
0.50 Molybdenum Electric Steel	(1650°F.—Air Cool (1200°F.—168 hrs. draw	61,050	81,710	29.0	176
0.15% Carbon (2) 5.10% Chrome	1600°F.—Furnace Cool	37,400	64,100	36.2	163
Electric Steel	(1600°F.—Furnace Cool ...... (1200°F.—168 hrs. draw	27,800	63,800	37.5	149
0.15% Carbon (4) 5.17% Chrome	1600°F.—Furnace Cool	45,600	77,400	31.0	170
0.50% Molybdenum Electric Steel	1600°F.—Furnace Cool ...... 1200°F.—168 hrs. draw	33,300	75,200	34.0	156

NOTE: (1) Steel Maker No. 1, (2) Steel Maker No. 2, (3) Steel Maker No. 3, (4) Steel Maker No. 4.

850 degrees Fahr. The National Tube Co. specimens discussed in this paper represent commercial lots of steel obtained from four different makers of high grade alloy steels in the United States, and we believe that this makes the results of particular importance from a design standpoint. We could not recommend 0.50 per cent molybdenum steel for use in refinery service at 1000 degrees Fahr. metal temperature for a stress of 9225 pounds (Table VI, author's paper) when we know that the tubing in service will gradually decrease in resistance to stress, and that its flow characteristics will accelerate as the softening develops in service. We would, however, feel justified in recommending a steel of this composition for a creep stress of 4750 pounds, since we feel the specimen subjected to testing approximates the properties which will be produced in the tube under operating conditions. It should also be pointed out that the creep tests discussed in this paper were only carried out for a period of 500 hours, and that if these tests had been extended for longer periods it is possible that the flow rate of the specimens which were not of a stable structure would have altered with time of test.

N. B. PILLING[2]: Perhaps the point has been dealt with somewhere in this paper, but I just wanted to make sure whether the inference from one of the charts shown on the screen is true—that this method of judging materials at high temperature might possibly be far from safe at much higher temperatures than 1000 degrees Fahr. Have you made any tests in this direction?

[2]Metallurgist, International Nickel Co., N. Y. C.

Authors' Closure (By A. E. White)

In reply to Mr. Pilling's question as to the possibilities of the time-yield test for indicating the creep characteristics at temperatures greater than 1000 degrees Fahr., the work herein reported was done primarily in the interests of certain power plants and they were not, at the moment, interested in temperatures greater than 1000 degrees Fahr. No work, therefore, was done at temperatures higher than 1000 degrees Fahr.

In regard to Mr. Bull's comments, I believe that in his second paragraph he gives in an excellent manner some of the existing variables which must be taken into consideration in creep work. It has been brought out in other papers, both by the authors and others, that there are a large number of variables which may affect creep. Among these may be mentioned chemical composition, heat treatment, grain size, both inherent and that produced through heat treatment, ingot size, degree of deformation during hot fabrication, the nature and extent of previous deformation, and method of manufacture.

Mr. Bull also points out that a large number of alloys are being made in many of our foundries today. Possibly as work goes on and more is found out about the properties of these various alloys, the number now being made may be greatly reduced. I think Mr. Bull will agree that if this number can be reduced it will be a great advantage to the producer. As a representative of consuming interests, I can also say it would be to their advantage. It is confusing to have claims presented concerning the advantages of certain products with little data given other than that of chemical composition.

In regard to the comments by Mr. Dixon, what has been said in reply to Mr. Pilling, I believe, will hold true in Mr. Dixon's case. We limited our temperature to 1000 degrees Fahr. I grant fully that it could go higher. Possibly the company with which Mr. Dixon is associated will sponsor work to carry it to a higher temperature. I think Mr. Dixon's criticism of creep is one which would have been more nearly true a few years ago than at the present time. I feel that most of the high temperature testing laboratories are today operating under very carefully controlled conditions.

Dr. Hatfield has stressed an important point in his statement as to what it is desired to procure from the creep test or from the time-yield test. We must not think of these tests in terms of the tests themselves, but rather—what is the desired result? We have found that some people are thinking of using the time-yield test for their engineering data. Others are using the test to determine the rate of creep at 1 per cent per 100,000 hours. Because of these two points of view, this comparison was undertaken.

Dr. Norton's conclusions concerning Fig. 2 are not clear to the authors. The elongation scale is plotted in terms of 40 millionths of an inch. In the original, the smallest dimension was 1 millionth of an inch, but for purposes of reproduction, the intervening lines were omitted. Likewise, the stress scale was divided into units of 100 pounds, but for reproduction, the intervening lines were again omitted.

Likewise, Dr. Norton's questions concerning Fig. 10 are not clear. This figure is a diagrammatic sketch of what we believe to be the variation in the relationship between Hatfield's time-yield and the creep characteristics with

temperature. The position of various steels on this chart is greatly influenced by their chemical composition, as we have attempted to show. For example, plain carbon steels at 800 degrees Fahr. fall to the left of the point of intersection, while the same steels at 1000 degrees Fahr. fall to the right.

In reply to Mr. McKinney, we agree that the time-yield results may be greatly influenced by such factors as the initial heat treatment of the steel. But we also feel that the creep characteristics may also be greatly changed. The question is whether or not the results from these two types of tests are changed an equal amount. If the heat treatment is such that an unstable structure is produced which tends to revert to a stable form during a 1000-hour testing period under stress, then it is evident that the results from the two tests would be influenced to a different degree.

The authors wish to express their appreciation for the full and comprehensive discussion of the paper written by E. C. Wright. His discussion points out the importance of the effects of heat treatment on the resulting creep properties. This is a matter which, in the early days of creep testing, was, to a considerable extent, overlooked by those who furnished creep data. The discussion by Mr. Wright goes beyond the mere analysis of the influence of heat treatment on creep properties as normally considered and construed. It discusses the effect and influence of a heat treatment which is given to steel for the purpose of producing it in as stable a state as possible. Yet, it is questionable if even the ultimate in stability has been produced by a treatment consisting of a normalizing or annealing operation followed by a draw at 1200 degrees Fahr. for 168 hours. Such a statement is made for the reason that, from work recently done in the University of Michigan laboratories, we have found that the spheroidized state is in itself a transition state, for with the temperature and time conditions suitable the carbides appear to form into parallel bands adjacent to grain boundaries, rather than remain as scattered aggregates in the crystals.

For instance, if the ½ per cent molybdenum steel mentioned in Mr. Wright's discussion would spheroidize at 1000 degrees Fahr. during its normal operating life, then it should be given a spheroidizing treatment preceding the test to determine its rate of creep. Also, the creep stress thus obtained—namely, 4750 pounds—should be used as the basis for engineering calculations. It is questionable, however, if that stress, in place of the 9225 pounds, also mentioned, should be used until it is demonstrated, beyond any doubt, that the steel during its normal life would spheroidize. What is true at 1000 degrees Fahr. is true with increasing force at lower temperatures. That is, the time to spheroidize at lower temperatures, such as 900 or 800 degrees Fahr., would doubtless run into hundreds of years, to say the least, so that the effects of spheroidization during the normal life of the steel would be nil.

The issue raised by Mr. Wright should be carefully considered. At temperatures above 1100 degrees Fahr. the advisability of making tests only on samples in a spheroidized condition should be carefully considered. At temperatures below 1000 degrees Fahr., however, there appears to be no need to conduct creep tests on only spheroidized stock as it is questionable, in fact, quite improbable, that the steels in service would be held for a sufficient length of time at the given temperatures to spheroidize.

THE PRESENT STATUS OF AGE-HARDENING

By Richards H. Harrington

Abstract

1. The theoretical aspects of age-hardening are briefly outlined for three general types: (a) simple precipitation age-hardening, (b) simple lattice-strain age-hardening and (c) complications due to allotropy.

2. Practical significance of age-hardening alloys at elevated temperatures: (a) for tensile strength and (b) for spring properties.

3. Methods of field search: (a) modifications of analyses already known and tested, (b) brief method of survey of ternary equilibrium diagrams.

4. The unique behavior of cobalt in age-hardening alloys.

Introduction

SATISFACTORY systems for classification of equilibrium diagrams have been developed. Some of the fundamental principles have been well-grounded and advanced to a degree of satisfactory practical application. The development of metallurgy in the future will probably be accomplished largely by detailed research within each of the main types of alloy combinations. The age-hardening type of alloy is certain to yield many new combinations, each characterized by certain specific practical applications.

The term "age-hardening" seems to have originated from the ability of certain alloys in quasi metastable states of equilibria to harden with the passing of time at ordinary so-called "room temperatures." The term has become broad in scope to include also those precipitation reactions that are more efficiently developed by the use of other temperatures.

I. Theories Involved in a Study of Age-Hardening

A. Simple Precipitation Hardening

The iron-tungsten system, as developed by Sykes,[1] is shown in

[1]W. P. Sykes, "The Iron-Tungsten System," *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. LXXIII, 1926.

A paper presented before the Fifteenth Annual Convention of the society held in Detroit, October 2 to 6, 1933. The author is a member of the society and is research metallurgist, Research Laboratories of the General Electric Co., Schenectady, N. Y. Manuscript received July 1, 1933.

Fig. 1 as representative of a precipitation hardening reaction. Alloys ranging from about 8 to 32 per cent of tungsten are relatively soft when quenched from above the solution temperatures, and are markedly hardened by reheating to suitable temperatures below the solution line. According to the Fe-W diagram, this is due to a critical

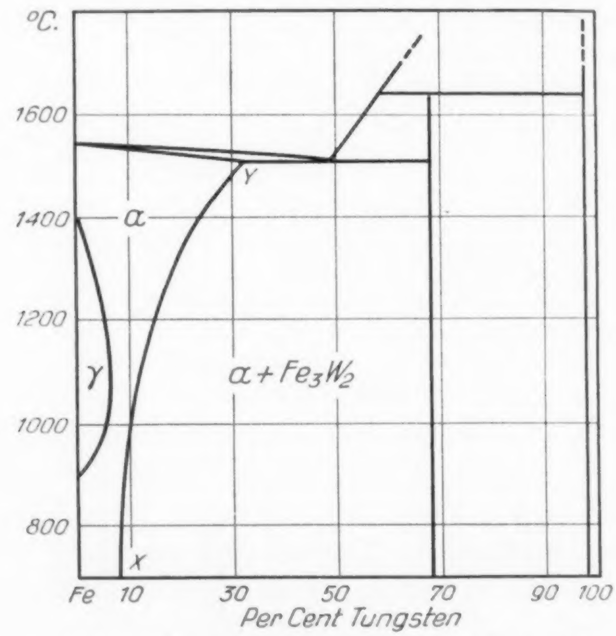


Fig. 1—Iron-Tungsten. XY is the Solution Line for the Age-Hardening Alloys.

dispersion precipitation[2] of Fe_3W_2. Thus an alloy of 15 per cent tungsten—85 per cent iron has a Brinell hardness of 147 as quenched in oil from 1200 degrees Cent. This increases to 230 when the alloy is reheated to 800 degrees Cent. and air-cooled. A similar age-hardening reaction applies, of course, to the low carbon steels (in particular) due to the solution of Fe_3C or carbon in alpha iron in the ranges of 0.02-0.10 per cent carbon (see Fig. 4).

B. Simple Lattice-Strain Hardening

Fig. 2 illustrates the age-hardening portion of the aluminum-copper system which is fundamental to most of the duralumin alloys. Work reported in 1929 by Gayler and Preston[3] led to their conclusions:

[2]The development of the critical dispersion theory of hardening is beautifully described in *Science of Metals*, by Jeffries and Archer, p. 390-401.

[3]"The Age-Hardening of Some Aluminum Alloys," by Marie L. V. Gayler and G. D. Preston, *Journal*, Institute of Metals, No. 1, 1929, Vol. XLI.

1. Changes in density and in lattice parameter during aging suggest that precipitation from the solid solution takes place.

2. X-ray analysis shows that the solid solution lattice is in a disturbed state which is gradually relieved by further aging at high temperature.

3. The increase in electrical resistance on aging corresponds to the distortion of the solid solution lattice.

4. It is inferred that the precipitation from solid solution involves two processes, (a) rejection of atoms of the solute from the

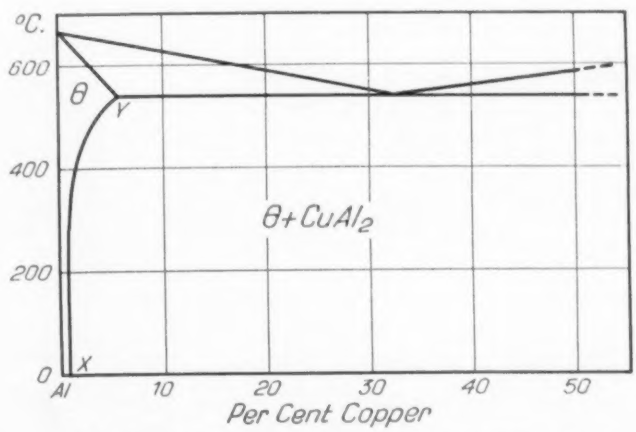


Fig. 2—Aluminum-Copper. XY is the Solution Line for the Age-Hardening Alloys.

solid solution lattice of the solvent, and (b) a "coagulation" process (corresponding to the formation of $CuAl_2$ particles, for example) which follows closely upon the first, and probably largely overlaps it.

In the discussion of this paper Dr. K. L. Meissner[3] gives tensile properties of a duralumin type of alloy that indicate a maximum strength for the alloy aged five days at room temperature. This alloy was then aged at elevated temperatures with a resulting minimum tensile strength after 100 degrees Cent. aging, followed by another maximum after about 150 degrees Cent. draw.

Further work by Gayler and Preston[4] in 1932 led to the following conclusions:

1. Age-hardening of duralumin at room temperature and at 200 degrees Cent. is due to some process, as yet undefined by either X-ray or microscopic examination, which takes place prior to actual precipitation of $CuAl_2$ or Mg_2Si from the aluminum lattice.

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[4]"The Age-Hardening of Some Aluminum Alloys of High Purity," by M. L. V. Gayler and G. D. Preston, *Journal*, Institute of Metals, Vol. XLVIII, No. 1, 1932.

2. It is suggested that Mg_2Si as well as $CuAl_2$ plays an important part in the process of age-hardening.

In discussion of this paper Dr. Sachs[4] states the following: At 150-200 degrees Cent. a hardening effect occurs at first, then disappears, and is finally followed by hardening of a different type. The characteristic duralumin hardening depends on effects which

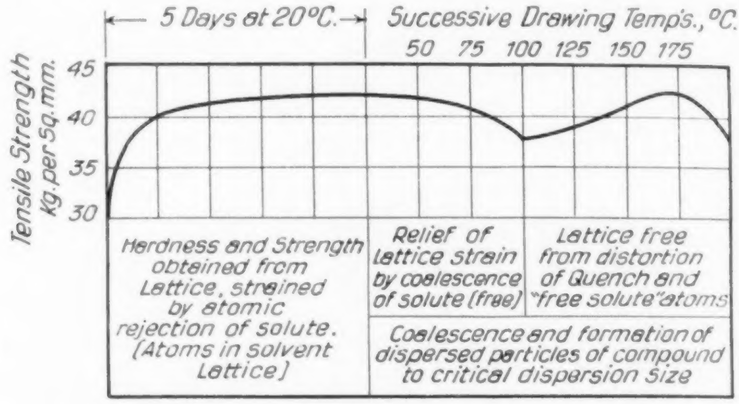

Fig. 3—Effect of Lattice-Strain Age-Hardening Duralumin.

occur within the solid solution. This is further clarified by the work of Schmid and Wassermann[5] and of Hengstenberg and Mark.[6] The lattice constant of duralumin after quenching shows a very slight falling off at the beginning of the process of aging at room temperature and shows no further changes during cold aging to maximum hardness. No $CuAl_2$ could be identified in the alloys aged at room temperature but there were indications that copper atoms were freed from solution and may seek special planes in the aluminum lattice. Aging at 150 degrees Cent. resulted in a hardness equivalent to that obtained by aging at room temperature but due to the identified existence of precipitated $CuAl_2$. Fig. 3 presents a summary of these theories.

It is seen, then, that simple lattice-strain may result from atomic precipitation of solute atoms from the solid solution lattice. It has been noted during the study of about 100 age-hardening alloys that many, after quenching from high temperatures, show an increase in hardness at low drawing temperatures followed by a minimum hardness and finally the maximum precipitation hardening at a higher

[5]E. Schmid and Wassermann, "Tempering of Duralumin. X-Ray Studies," *Metallwirtschaft*, Vol. 9, 1930, p. 421-5.

[6]J. Hengstenberg and H. Mark, "X-Ray Studies of Lattice Structures in Light Metals," *Zeitschrift für Electrochemie und Angewandte Physikalische Chemie*, Vol. 37, 1931, p. 524-28.

temperature. In many alloys the precipitation of a true compound begins to take place immediately upon heating and masks any tendency to produce simple lattice-strain hardening.

C. *Complications Due to Allotropy*

The Fe-C system (equilibrium diagram shown in Fig. 4) illustrates very well the complications added by allotropy to otherwise simple precipitation hardening reactions. Because of this complication, the system, though one of the oldest known to metallurgists, is still the subject of a great deal of active research.

Only comparatively recently was it generally recognized that austenite, retained in quenched steels, could cause actual age-harden-

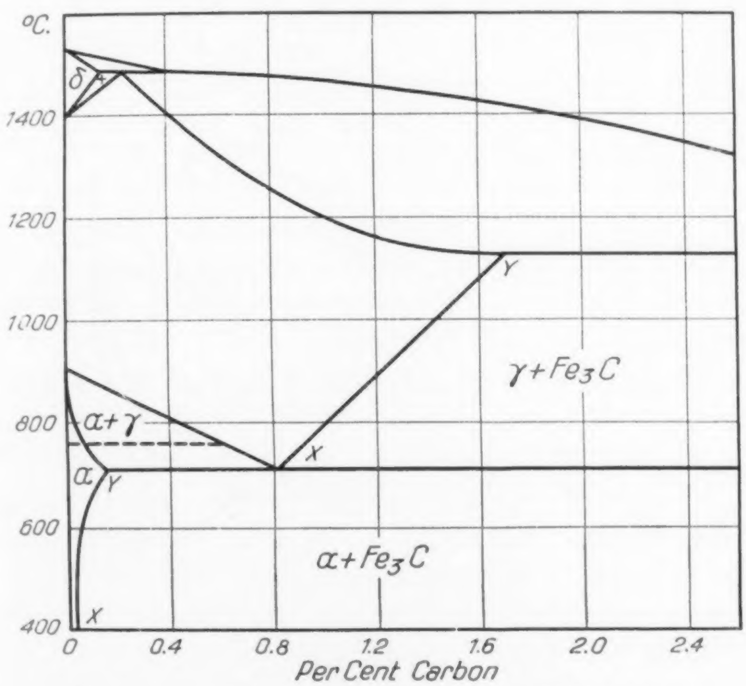

Fig. 4—Iron-Carbon System. XY Indicates Solution Line.

ing of these quenched steels due to the completion of the A_3 allotropic change during low temperature draws (up to 180 degrees Cent.). This additional hardening may be variously ascribed to lattice strain-hardening due to atomic carbon free from solution, or to critically dispersed Fe_3C, or to both. Still more recent has been the recognition of the tetragonal form of martensite, intermediate to austenite and cubic martensite.

According to the diagram (the cementite solution line above

700 degrees Cent.), hypereutectoid alloys should be age-hardening but any effect of this type of Fe_3C precipitation is largely masked by the allotropic transformation at 700 degrees Cent. that always tends to take place to considerable degree in these quenched alloys. These effects are, however, subject to study.[7]

D. Other Causes of Age-Hardening Effects

Some of these are as follows:

1. Gas reactions within a metal lattice (of which blue-brittleness may be an example).

2. A single solid solution phase in a quenched alloy may, upon reheating, transform into one or two solution phases, of which either one or both may be harder than the original.

3. Cold-worked metals and alloys may show age-hardening phenomena due to recrystallization to a fine grain size or to especially efficient precipitation of a new phase along the lattice planes of maximum slip or deformation.

II. Age-Hardening Alloys for Use at Elevated Temperatures. Specific Requirements

A. For Tensile Properties

If the iron-tungsten alloy (15 tungsten-85 iron as cited above) could be modified by the addition of other elements to strengthen the solid solution matrix and also decrease the solubility of the precipitating compound without appreciably lowering the critical dispersion temperature, the resulting alloy might have excellent tensile properties for uses at high temperatures.

If the temperature range between the critical dispersion drawing temperature and the solution temperature is relatively great, it could be expected that grain growth of the precipitant would be relatively small in degree and slow at temperatures slightly above the critical temperature, and such an alloy might be safely used at temperatures even above the critical aging temperature without exceeding the required creep resistance values.

To the 15 per cent tungsten-85 per cent iron alloy cobalt was added, keeping the ratio of tungsten to iron about 15 to 85. The resulting alloy had a composition of 15 per cent cobalt-12 per cent

[7]R. H. Harrington, "Effect of Prequench Treatments on the Hardness, Body, and Structure of Hardened Tool Steel," Transactions, American Society for Steel Treating, Vol. 18, 1930, p. 404-423.

tungsten-73 per cent iron. This alloy has a maximum age-hardness of 435 Brinell after drawing at 700 degrees Cent. as compared to 230 Brinell for the 15 per cent tungsten-85 per cent iron alloy after drawing at 800 degrees Cent. The tensile strength of the cobalt-tungsten-iron alloy is 107,000 pounds per square inch at 800 degrees Cent. as compared to 32,000 pounds per square inch for the tungsten-iron alloy at 800 degrees Cent. The data are given in Table I.

B. For Spring Properties

If an alloy could be obtained with a matrix of high damping capacity, keyed by a critically dispersed phase to resist actual deformation, an ideal spring material might result. It would seem that many metals in a nearly pure state might well serve for the matrix. If, then, to a binary age-hardening alloy, a third element could be added to nearly completely precipitate as a compound with one of the elements of the original binary alloy, the resulting ternary alloy would undoubtedly have interesting properties. Such an alloy could be used efficiently only at temperatures below that of critical dispersion. At temperatures just above the critical value, the grain size of the precipitant would tend to grow and, as the solution temperature is approached, more of the precipitant would dissolve in the matrix and probably reduce the damping qualities of the alloy as well as the hardness and resistance to deformation.

It must be recognized also that the time factor at any one temperature must be thoroughly considered for each special application.

A binary alloy of 99.6 per cent copper-0.4 per cent beryllium, melted in a high-frequency vacuum furnace, had a maximum hardness of 53 Brinell when aged at 200 degrees Cent. To this alloy was added some cobalt to give a composition of 2.6 per cent cobalt-0.4 per cent beryllium-97 per cent copper.[8] This alloy was also melted in a high frequency vacuum furnace. Quenched from 900 degrees Cent. into water, the conductivity was 20 per cent of that of copper. After aging at 500-600 degrees Cent. to a Brinell hardness of 180-200 the conductivity increased to 43 per cent. It is, therefore, assumed that most of the cobalt and beryllium (and probably some copper) have been precipitated, leaving only relatively small amounts of these elements in solid solution in the copper matrix.

The tensile properties of this ternary alloy are of considerable

[8]Otto Dahl. Patent No. 1,847,929. Assigned to the General Electric Co.

Table I
Aging Characteristics and Tensile Properties

No. of Alloy	Composition	Forge? Temp.	Brinell as Cast	Hardness as Heat Treated	Drawn at—Degrees Cent. 200	300	400	500	600	700	800	900
56	85 Fe-15 W	Yes 1100°C	149	1200°C Oil 147	145	147 Y	149 b	147 b	156 s	187 s	230 — s	202 s
57	15 Co-73 Fe-12 W	Yes 1100°C	300	1200°C Oil 286	277	286 Y	293 Y	321 b	351 g	435 — d	293 d	228 d

No. of Alloy	Composition	Tested at Room Temp. T. S. ×1000	Y. P. ×1000	E. L. ×1000	El. %	R. A. %	Tested at 600°C $^{+20°}_{-0°}$ T. S. ×1000	Y. P. ×1000	E. L. ×1000	El. %	R. A. %	Tested at 800°C $^{+25°}_{-0°}$ T. S. ×1000	Y. P. ×1000	E. L. ×1000	El. %	R. A. %
56	85 Fe-15 W Forged & Aged	111	98	65	6	6	52	52	50	25	61	32	30	28	31	80
57	15 Co-73 Fe-12 W Forged & Aged											107	104	83	3	3

Table II
Aging Characteristics and Tensile Properties

No. of Alloy	Composition	Forge? Temp.	As Cast	As Heat Treated	Drawn at—Degrees Cent. 200	300	400	500	600	700	800	900
54	99.6 Cu-0.4 Be Hardness	Yes 800°C	38 Brinell	900°C H_2O 50 Brinell	53 —	50 g	42 g	42 s	38 s	34 s	47 s	42 s
	Conductivity		38%	38.8%	38.9%							
55	97 Cu-0.4 Be-2.6 Co Hardness	Yes 800°C	80 Brinell	900°C H_2O 69 Brinell	69	74 g	80 g	143 s	200 — s	92 s	80 s	80 s
	Conductivity		26.3%	19.6%					43%			

No. of Alloy	Composition	Tested at Room Temp. T. S. ×1000	Y. P. ×1000	E. L. ×1000	El. %	R. A. %	Tested at 350°C T. S. ×1000	Y. P. ×1000	E. L. ×1000	El. %	R. A. %	Tested at 475°C T. S. ×1000	Y. P. ×1000	E. L. ×1000	El. %	R. A. %
55	97 Cu:0.4 Be-2.6 Co	90.2	90.2	61	20.3	23.6	67.5	67.5	61	4	4	56	56	40	0	0

interest: at room temperature an elastic limit of 61,000 pounds per square inch with a tensile strength of 90,000 pounds per square inch; at 350 degrees Cent. an elastic limit of 61,000 pounds per square inch and a tensile strength of 67,500 pounds per square inch. Age-hardening data and tensile properties are shown in Table II. The elastic limits as given are only approximate values.

III. Methods of Field Search for Age-Hardening Alloys

Procedure

Unless otherwise noted, 200-gram melts of the alloys were made in small induction furnaces and poured into graphite molds. Hardness tests were made on the cast alloy. Samples were then quenched from high temperatures, usually 950 degrees Cent. (also, often, 1200 degrees Cent.), and then aged at successively higher temperatures. Hardness tests, either Rockwell or Brinell, were made after each heat treatment and all hardness tests expressed in Brinell values for ease of comparison.

Oxidation colors and scales were noted after each heat treatment and recorded in the hardness tables according to the following system:

s = straw	d = dark
y = yellow	g = gray
b = blue	S = badly scaled
br = brown	

Many of the alloys have been tested for forgeability and the forging temperatures noted. Hand hammering, machine rolling, and swaging have yielded this information. Some of the alloys have been magnetically tested and the results recorded as "magnetic, weak, or nonmagnetic." Following the tabulation of this general knowledge, specific alloys may indicate properties for special uses.

A. The First Method of Field Search

This is to vary the composition and the heat treatments of the alloys already known to be age-hardening and to employ additive elements in new compositions. One example of this method is the history of high speed steels following the discovery by Taylor and White that heat treated chromium-tungsten steels retained useful hardness at temperatures up to about 600 degrees Cent. Research has since developed many tool alloys by adding cobalt and vanadium

Table III
Properties of Age-Hardening Tool Materials

No.	Composition	Forge? Temp.	As Cast	Hardness After Heat Treatment	Drawn at—Degrees Cent. 200	400	500	600	700	800	900
41	36 Co-8 Mo-5 Cr 1 C-50 Ni	Yes 1000°C	192 Brinell	950°C Oil Quench 225	230	230	235	230	202	207	202
42	36 Co-8 Mo-5 Cr 1 C-.4 V-49 Fe	Yes 1100°C	500 Brinell	950°C Oil Quench 650	670	713	817	620	550	450	[illegible]
43	Fe Co W	Yes 1100°C	480 Brinell	1100°C Oil Quench 444	465	555	740	745	600	495	[illegible]

and substituting molybdenum for tungsten. One of the latest developments is alloy No. 548 consisting chiefly of cobalt, tungsten, and iron[9]—eliminating carbon from the high speed steel analysis.

Three alloys of marked chemical similarity in composition, but of widely different metallurgical properties, are described in Table III. Alloy 42 age hardens at 500 degrees Cent. to about 73 Rockwell C or 88A. Although its "red-hardness" is at a comparatively low temperature, 500-550 degrees Cent., its great hardness and ability to hold an exceedingly keen edge have adapted it to certain finish and semifinish cutting operations. Alloy 41 differs from 42 only in the substitution of nickel for iron. Alloy 41 does age harden but at a much lower hardness and is of no use as a tool material. It is still one of the mysteries that elements of the same chemical classification and of considerable similarity in atomic structure should differ so widely in metallurgical properties. Alloy 43 is one of the "548" class of compositions and its high "red-heat hardness," extending even above 700 degrees Cent., is some indication of its outstanding efficiency in heavier cutting operations. Comparing alloy 42 to 43 one notes again that the presence of carbon in this analysis tends to lower the temperature of maximum age-hardness.

It is well known that nickel-chromium alloys of 70-30 and 80-20 per cent compositions[10, 11] have excellent high temperature properties. Thus the 70 per cent nickel-30 per cent chromium alloy is forgeable and has a tensile strength of 43,000 pounds per square inch at 800 degrees Cent. If additions could be made to such an alloy to age harden it at temperatures from 600 to 900 degrees Cent.,

[9]Dr. Zay Jeffries and W. P. Sykes, "Alloy 548—A New Metal Cutting Alloy," *Metal Progress*, February, 1933.

[10]W. Rosenhain and C. H. M. Jenkins, "Some Alloys for Use at High Temperatures," *Journal*, Iron and Steel Institute, Vol. CXXI, No. 1 for 1930.

[11]Jenkins, Tapsell, Austin and Rees, "Some Alloys for Use at High Temperatures," *Journal*, Iron and Steel Institute, Vol. CXXI, No. 1 for 1930.

Table IV
Brinell Hardness, Temper Colors, and Forging Properties of Tungsten-Chromium-Nickel Alloys

No.	Composition	Forge? Temp.	As Cast	950 Oil	Drawn at—Degrees Cent. 200	400	600	700	800	900
1	2.5 W-78 Ni-19.5 Cr	? 1200°C	137 Brinell	168	180	163	166	172	163	166
							b	y	g	d
2	10 W-72 Ni-18 Cr	No	153 Brinell	185	186	180	180	185	179	180
							b	y	g	d
3	20 W-64 Ni-16 Cr	No	170 Brinell	194	199	205	205	212	196	207
							b	y	g	d
4	30 W-56 Ni-14 Cr	No	217 Brinell	212	217	217	217	207	215	223
							b	y	g	d
5	2.5 W-68 Ni-29.5 Cr	? 1300°C	153	183	196	202	196	205	202	196
							b	y	g	d
6	10 W-63 Ni-27 Cr	? 1300°C	179	196	210	215	215	235	207	207
							b	y	g	d

it might be expected that the tensile properties in this temperature range could be improved. Although the nickel-tungsten system is apparently not yet accurately known, data do exist that indicate an age-hardening series of alloys from 10-34 per cent tungsten. The results of an age-hardening study of nickel-chromium-tungsten alloys are tabulated in Table IV. These alloys do not appear to be forgeable and, while the tensile strengths of some of them were comparable to the nickel-chromium alloys without tungsten, they do not appear to be of practical use. It was noted, however, that most of these alloys developed two peaks of maximum age-hardness—one from 200-400 degrees Cent. and the second about 700 degrees Cent. There is, at present, no other explanation than that these alloys can be classed with duralumin as showing one peak hardness due to solvent lattice strain caused by liberated solute atoms and a second peak hardness due, probably, to critically dispersed compound.

B. *The Second Method of Field Search*

This is to draw up certain promising binary and ternary systems and to study their possibilities in a briefly qualitative way. In the case of ternary systems this may often be done to advantage by crossing the composition diagram with a single straight line, maintaining the ratio of two of the elements a constant and varying the third. From such simple alloys of specific interest may be developed more complex alloys with various additional elements. Thus the nickel-

iron-titanium system, as it is incompletely known today, is shown in Fig. 5 and the results of a brief age-hardening study of the indicated alloys are given in Table V. Due to high cost of pure titanium and some indications that the high titanium content alloys are not of practical value, the nickel-titanium and iron-titanium systems have

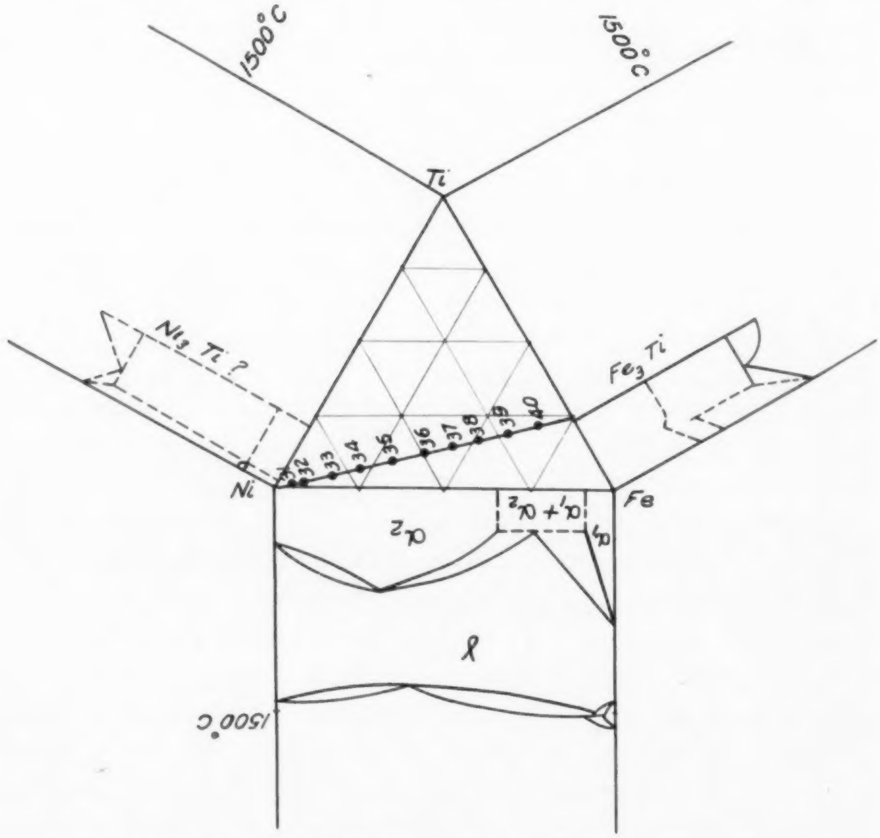

Fig. 5—Nickel-Iron-Titanium System, as Incompletely Known Today.

apparently not been thoroughly investigated. Alloys containing up to 5 per cent titanium are forgeable and alloy 33 definitely age-hardens at 700 degrees Cent. The high nickel content appeared to give alloys 31, 32, 33 considerable resistance to oxidation at temperatures below 700 degrees Cent. It seems probable that a ternary compound is formed in the neighborhood of alloys 36 and 37. The alloys of generally useful properties appear to be those of high nickel content and containing from 2 to 5 per cent titanium. The addition of chromium to these alloys has been described by Dr. Kroll.[12] Austin

[12]Dr. W. Kroll, "Hardenable Titanium Alloys," *Metallwirtschaft*, Vol. 9, Dec. 19, 1930, p. 1043-45.

and Halliwell[13] have described the Konel type of nickel-iron-cobalt alloys with additions of titanium and chromium. The age-hardening characteristics of these alloys appear to be dependent upon a

Table V

Properties of Alloys of the Ni-Fe-Ti Ternary System

No.	Composition	Forge? Temp.	As Cast	950 Oil	Drawn at—Degrees Cent. 200	400	500	600	700	800	900	Magnetic As Cast?
31	1.25 Ti-3.75 Fe-95 Ni	Yes 1150°-1200°C	113 Brinell	105	108	107	108	113	112	108	106	Strongly
32	2.5 Ti-7.5 Fe-90 Ni	Yes 1150°-1200°C	143 Brinell	134	137	143	143	137	145	143	134	Strongly
33	5 Ti-15 Fe-80 Ni	Yes 1200°C	277 Brinell	260	262	262	262	260	311	302	277	Strongly
34	7.5 Ti-22.5 Fe-70 Ni	?	340 Brinell	351	351	351	351	351	390	395	364	Weakly
35	10 Ti-30 Fe-60 Ni	No	430 Brinell	387	380	387	395	375	490	440	387	Weakly
36	12.5 Ti-37.5 Fe-50 Ni	No	460 Brinell	420	444	440	450	500	512	512	555	No
37	15 Ti-45 Fe-40 Ni	No	550 Brinell	450	430	402	460	450	512	578	500	No
38	17.5 Ti-52.5 Fe-30 Ni	No	550 Brinell	444	460	440	532	532	578	683	540	Slightly
39	20 Ti-60 Fe-20 Ni	No	420 Brinell	460	440	387	500	450	460	500	532	Weakly

precipitation reaction involving titanium specifically and qualified by the other elements.

With two exceptions the alloys in Table V also show two age-hardening peaks. An explanation is not so readily forthcoming, however, as these alloys probably have their aging reactions complicated

[13]Austin and Halliwell, "Some Developments in High Temperature Alloys of the Ni-Co-Fe System," American Institute of Mining and Metallurgical Engineers, Technical Publication No. 430, 1931.

Table VI
Effect of Cobalt on Some Age-Hardening Alloys

No.	Composition	Forge? Temp.	As Cast	Hardness As Heat Treated	200	300	400	500	600	700	800	900	Maximum Increase Hardness
								Drawn at—Degrees Cent.					
51	98 Ni-2 Be	Yes 1000°C	235 Brinell	950°C H_2O 140	150	165	165	375	311	262	202	145	235
52	95 Ni-5 Be	Yes 1000°C	490 Brinell	950°C H_2O 302	302	322	340	480	480	340	286	296	178
53	10 Co-88 Ni-2 Be	Yes 1000°C	196 Brinell	950°C H_2O 185	185	187	192	293	387	269	217	179	202
							y	y	b	b	y	g	247a
54	99.6 Cu-.4 Be	Yes 800°C	38 Brinell	900°C H_2O 50	53	50	42	42	38	34	47	42	3
					—	g	g	s	s	s	s	s	
55	97 Cu-2.6 Co-.4 Be	Yes 800°C	80 Brinell	900°C H_2O 69	69	74	80	143	200	92	80	80	109
						g	g	s	s	s	s	s	128a
56	85 Fe-15 W	Yes 1100°C	149 Brinell	1200°C Oil 147	145	147	149	147	156	187	230	202	83
						y	b	b	s	s	s	s	
57	15 Co-73 Fe-12 W	Yes 1100°C	300 Brinell	1200°C Oil 286	277	286	293	321	351	435	293	228	149
						y	y	b	g	d	d	d	288a
58	97 Fe-3 Be	?	265 Brinell	950°C H_2O 212	223	248	277	364	418	321	277	277	206
						y	b	b	y	b	d	d	
59	15 Co-82.5 Fe-2.5 Be	?	320 Brinell	950°C H_2O 255	255	277	320	387	460	387	340	302	205
						y	b	b	y	b	d	d	248a
60	52 Ni-36 Co-9 Fe-3 Ti	Yes 1100°C	202 Brinell	950°C Air 150			150	163	207	286	185	143	136
							y	y	y	g	g	g	152a

by the allotropy of nickel between 350-400 degrees Cent. This transformation is extended across the nickel-iron diagram to merge with that of iron at 25 per cent iron.

IV. Unique Behavior of Cobalt in Age-Hardening Alloys

The addition of cobalt to several alloys (as in alloys 42 and 43 of Table III) has resulted in a marked increase in the age-hardness of these compositions. It became of interest to study the effect of cobalt additions to a wide variety of age-hardening alloy compositions. Table VI shows the results of tests on ten of these alloys representing

five pairs of comparable systems. In the last column (to the right) is given the difference in hardness between the "as-quenched" and the maximum age-hardened states of each alloy. The figures given with the small letter "a" indicate the differences between the maximum age-hardness of the cobalt alloys and the "as-quenched" hardness of the similar alloys without cobalt.

The nickel-beryllium alloys (51 and 52) were melted in a hydrogen furnace. These show exceptional age-hardening properties and excellent resistance to oxidation at high temperatures. The percentage increase in age-hardness decreases with increase in beryllium content. Increasing beryllium content raises the temperature of maximum age-hardness. It is well-assured, then, that this range of the nickel-beryllium system includes a solution line similar in character to the line xy in Fig. 1. Alloy 53 shows the effect of adding cobalt to the composition of Alloy 51. The maximum age-hardness is greater and the temperature of age-hardening is increased by 100 degrees Cent.

Alloys 54 and 55 were melted in a vacuum induction furnace and show the effect of cobalt additions to copper-beryllium alloys of low beryllium content. (These alloys have been discussed in greater detail earlier in the paper. See Table II.) Here the cobalt addition accomplishes another remarkable effect—that of actually increasing the electrical conductivity to 45 per cent for the age-hardened alloy as compared to 38 per cent for the similar alloy lacking cobalt.

Alloys 56 and 57 illustrate the effect of cobalt when added to some iron-tungsten alloys. The 15 per cent tungsten-85 per cent iron alloy ages quite a bit at 800 degrees Cent., but the addition of cobalt more than doubles this increase in hardness while tending to lower somewhat the temperature of maximum aging. This is the only instance, among the alloys so far studied, in which the addition of cobalt appreciably lowered the aging temperature. In all the other alloys the effect of cobalt was generally to raise the aging temperature. (These alloys have been discussed in greater detail in Table I.)

Alloys 58 and 59 show the effect of adding cobalt to beryllium-iron alloys. The age-hardness was again increased by the cobalt addition. These alloys are not forgeable and, under compression, "cry" as does tin.

Alloy 60 may be considered as showing the increased age-hardness resulting from the addition of cobalt to Alloy 32 of Table V.

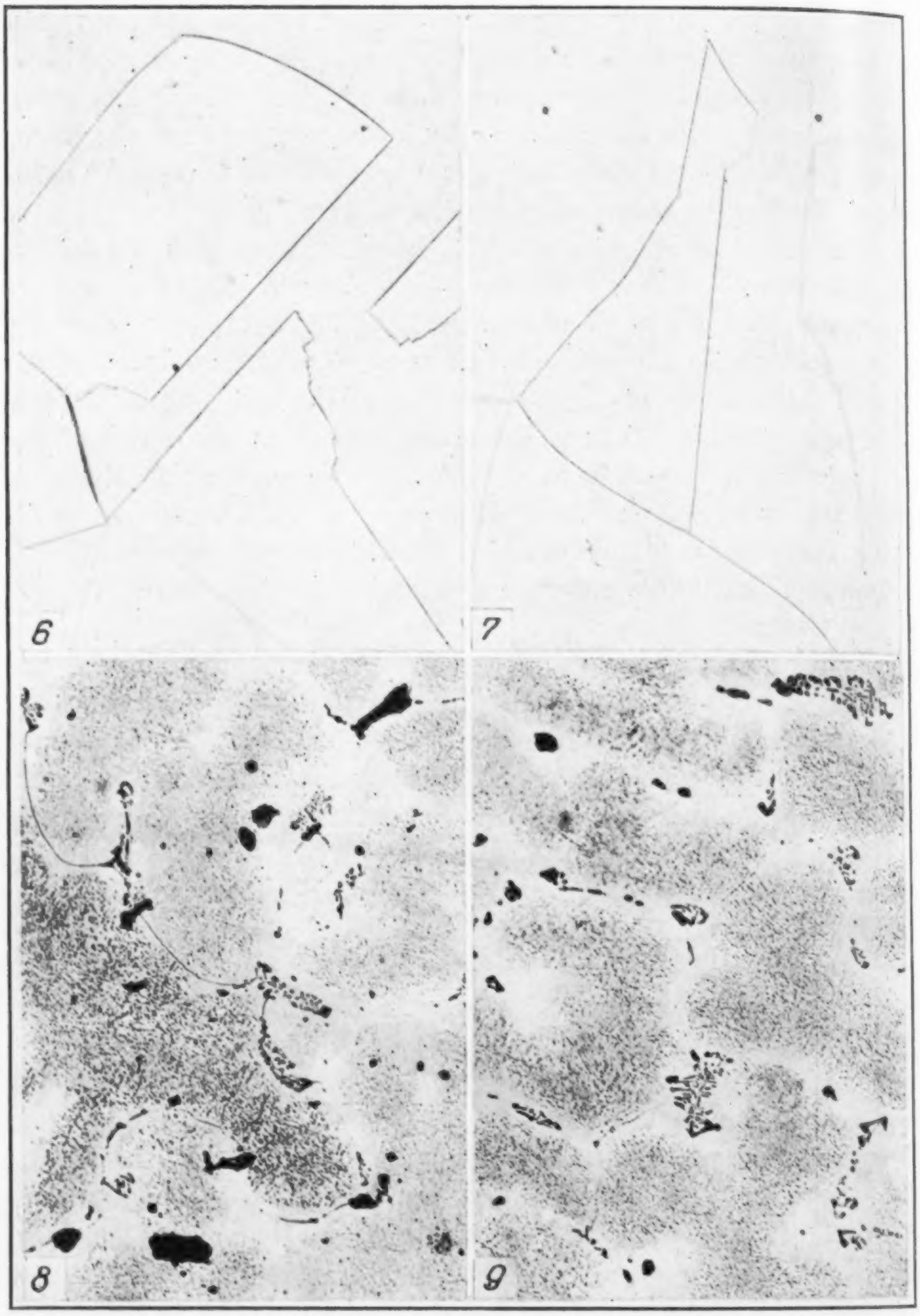

Fig. 6—Alloy 54, Containing 99.6 Copper, 0.4 Beryllium. Quenched in Water from 900 Degrees Cent. Note Unusual Steps in Grain Boundaries.
Fig. 7—Same as Fig. 6 with a 200 Degrees Cent. Draw.
Fig. 8—Alloy 55, Containing 2.6 Cobalt, 0.4 Beryllium, 97.5 Copper. Quenched in Water from 900 Degrees Cent.
Fig. 9—Same as Fig. 8 with a 600 Degrees Cent. Draw.

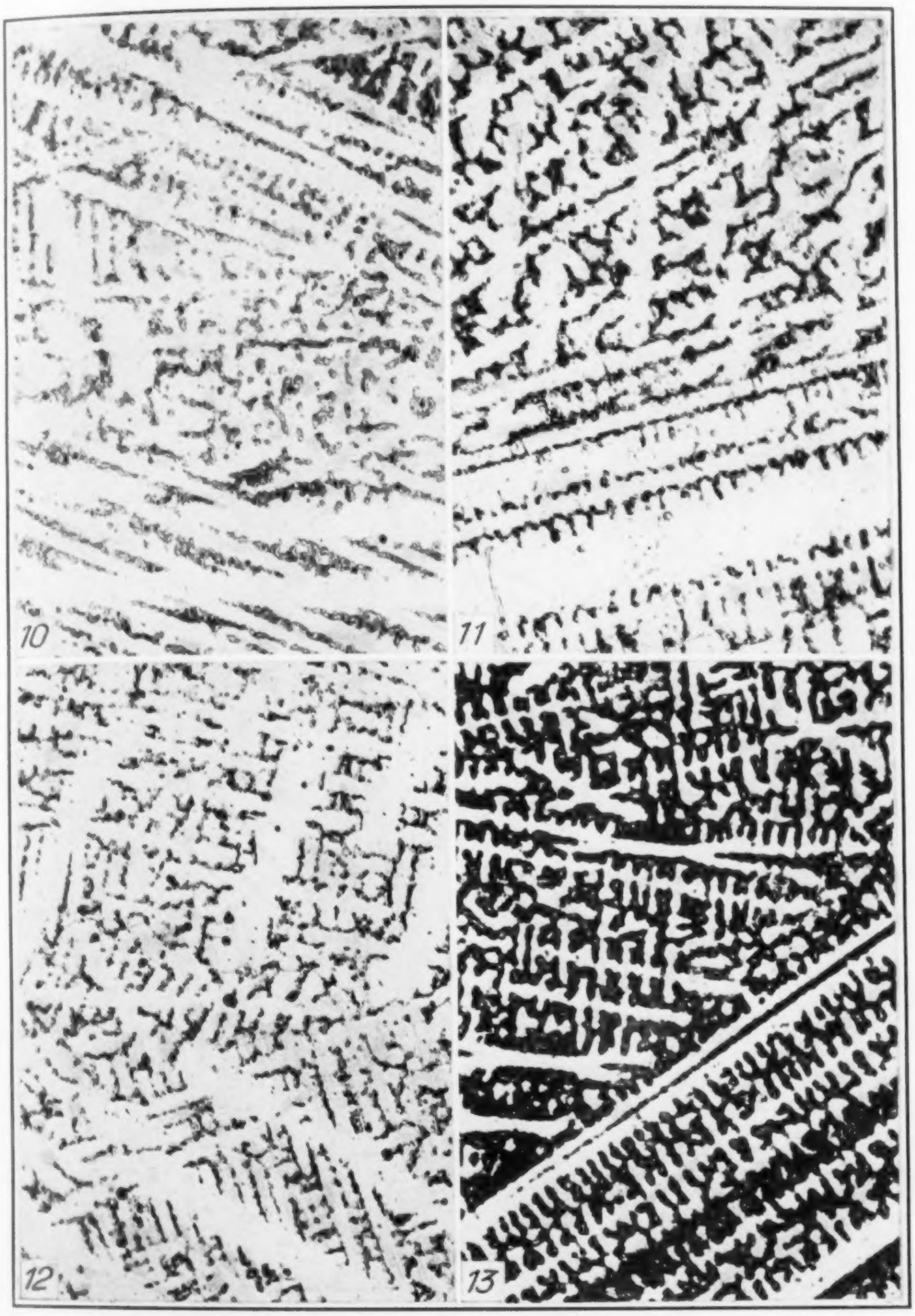

Fig. 10—Alloy 51, Containing 98 Nickel, 2 Beryllium. As-Cast.
Fig. 11—Same as Fig. 10 with 950 Degrees Cent. Water Quench.
Fig. 12—Same as Fig. 11 with 500 Degrees Cent. Draw.
Fig. 13—Alloy 53, Containing 88 Nickel, 2 Beryllium, 10 Cobalt. As-Cast.

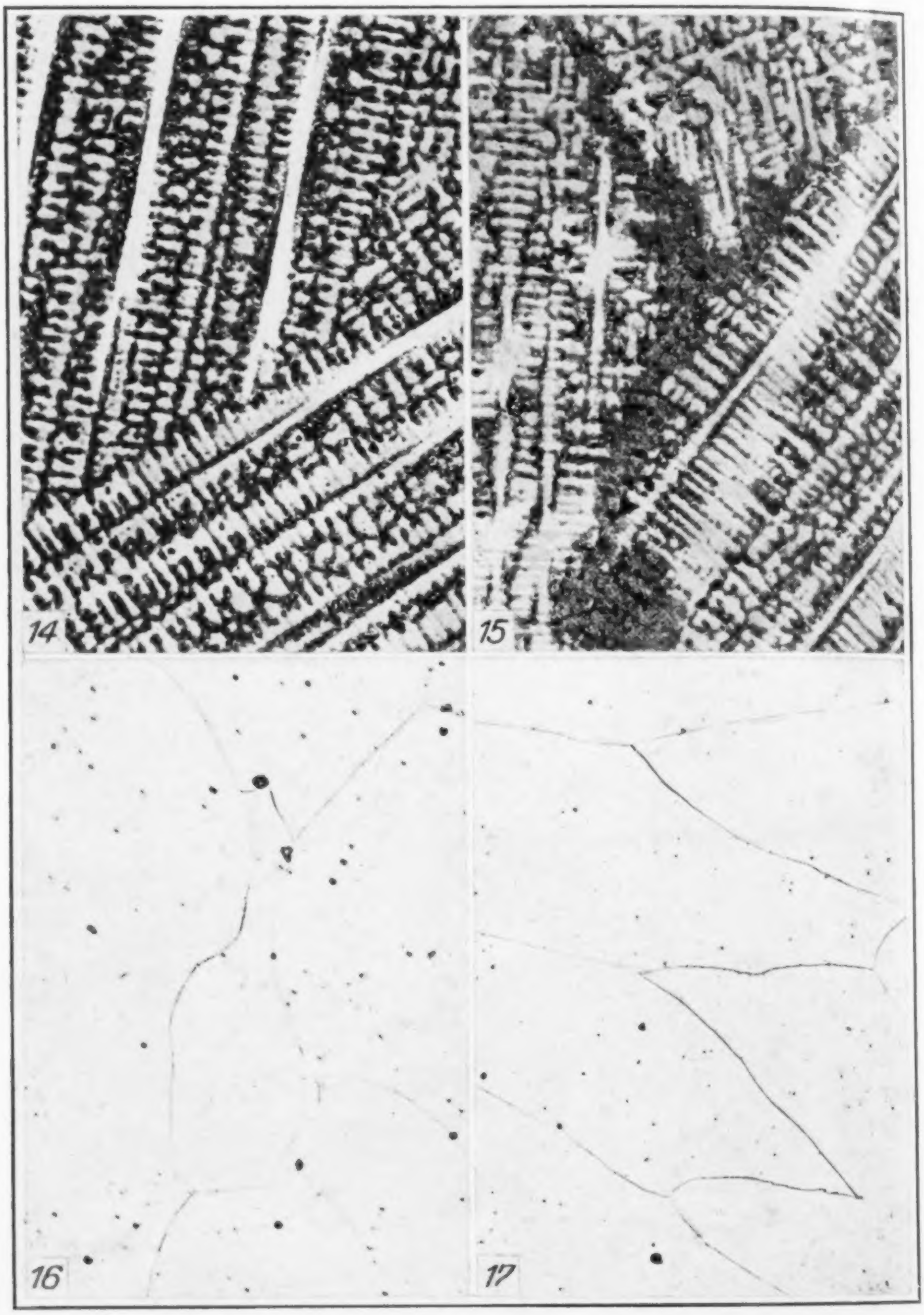

Fig. 14—Same as Fig. 13 Quenched in Water from 950 Degrees Cent.
Fig. 15—Same as Fig. 14 with a 600 Degrees Cent. Draw.
Fig. 16—Alloy 58, Containing 97 Iron, 3 Beryllium. As-Cast.
Fig. 17—Same as Fig. 16 with 950 Degrees Cent. Water Quench.

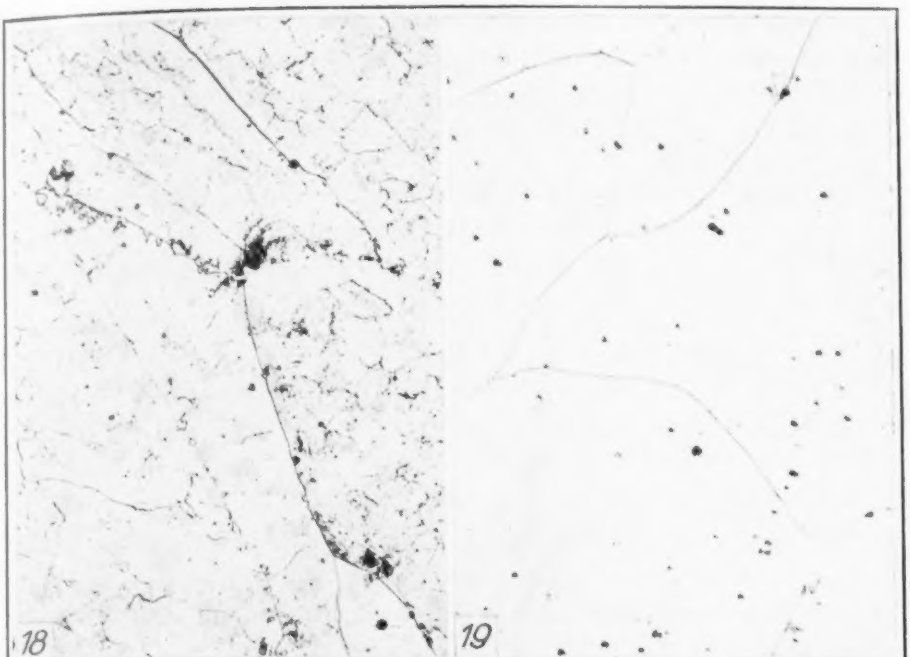

Fig. 18—Same as Fig. 17 with 600 Degrees Cent. Draw.
Fig. 19—Alloy 59, Containing 82.5 Iron, 2.5 Beryllium, 15 Cobalt. As-Cast.
Fig. 20—Same as Fig. 19 with 950 Degrees Cent. Quench in Water.
Fig. 21—Same as Fig. 20 with 600 Degrees Cent. Draw.

The general conclusion is that all of the alloys (studied in this research) had their age-hardness increased by cobalt additions.

Summary

1. Age-hardening (exclusive of phenomena dependent upon cold working) can now be briefly classified in three general types: (a) simple precipitation hardening, (b) simple lattice-strain age-hardening and (c) complications due to allotropy.

2. Practical significance of age-hardening alloys at elevated temperatures is suggested in relation to tensile properties and to spring properties.

3. High speed steel tools (and other alloys) of recent origin are logical modifications of analyses previously known and tested.

4. A brief qualifying search of ternary systems is described.

5. Cobalt, added to age-hardening alloys, acts in general to further increase the age-hardness and often to raise the temperature of maximum age-hardness.

6. Cobalt, added to age-hardening alloys, may act generally in two ways: (1) as a "desolvent," reducing the solubility of the precipitating constituent, at least part of the cobalt going into solid solution in the solvent lattice; (2) almost all of the cobalt, (if correctly added) enters into a ternary constituent with the precipitating agent, thus usually decreasing the solubility of this precipitant.

7. At the present time very few practical applications can be based on the knowledge of only the hardness variations in an alloy. Knowledge of certain other properties is very necessary to each specific use of any alloy. Yet, as a fuller understanding of the nature of age-hardening is achieved, it can be hoped that relations to other properties (such as electrical conductivity and magnetic properties) will become as corollaries to the general principles. It is intended in the future to make a careful study of certain other properties of the alloys reported herein.

Acknowledgments

The writer wishes to express his sincere appreciation of the advice given freely by members of the staff and he is especially grateful to Messrs. G. H. Howe, W. E. McKibben, and M. D. Collins who fabricated many of the alloys in high frequency and hydrogen muffle furnaces.

DISCUSSION

Written Discussion: By W. P. Wood, professor of metallurgical engineering, University of Michigan, Ann Arbor, Mich.

Dr. Harrington has given us a resumé of the present conceptions of age-hardening, various methods for the development of age-hardening alloys and the rather outstanding behavior of cobalt in causing the age-hardening phenomenon to appear in various alloys. This last point is of interest, since cobalt as a simple alloying element has not seemed as effective in many cases as other metals, such as nickel, chromium, molybdenum and others well known.

As I read the paper the question arose in my mind as to whether or not the various methods of hardening metals and alloys will be found in the near future to be all merely variations of the same thing. We have work-hardening, hardening by thermal means and age-hardening. It would be very helpful if this question of hardening could be boiled down to one line of attack and I feel that we are working in that direction.

Dr. Harrington's statements regarding the possibilities of age-hardening alloys for spring materials were of interest because they brought to mind some experiments which we have recently carried out upon helical springs made from high speed steel. The utility of such springs at elevated temperatures is entirely dependent upon the drawing operation. Unless this has been done in such a way as to produce what we may call critical dispersion the springs will show losses in loading capacity which make them no better than other simpler steels. In this same connection it would be of great interest to study the effects of the addition of third elements, such as nickel or beryllium to bronze, since at the present time bronze springs have little value above 100 degrees Fahr.

Written Discussion: By W. P. Sykes, Cleveland Wire Works, General Electric Co., Cleveland.

Dr. Harrington has presented a valuable outline of the hardening characteristics of many ternary alloys, and an interesting description of a plan for exploring an unknown system.

I wonder, however, if further investigation will not prove the absence of any fundamental differences between his types A and B? Such an accomplishment will necessitate the use of precise methods, such as measurement of electrical resistance and studies of X-ray diffraction patterns. To the best of my knowledge neither of these methods has been applied to the study of age-hardening in the iron-tungsten system or the iron-molybdenum system, which would belong to Type A by Dr. Harrington's classification.

In raising this question I wish to present some data of a preliminary nature, recently obtained at the laboratory of the Cleveland Wire Works, during the study of resistance changes which accompany age-hardening in alloys of the cobalt-tungsten system.

In this system, as recently described,[1] the solid solubility of tungsten in cobalt decreases from about 32 per cent at 1100 degrees Cent. to about 3 per cent at 500 degrees Cent., and age-hardening is brought about in the supersaturated solid solution by heating at a temperature as low as 500 degrees

[1]W. P. Sykes, "The Cobalt-Tungsten System," TRANSACTIONS, American Society for Steel Treating, Vol. XXI, No. 5, 1933, p. 385.

Cent. This system, it would appear, is quite comparable to the iron-tungsten system and for that reason the following data would seem to have a proper place in this discussion.

The changes in the electrical resistance accompanying aging at 500, 600, and 700 degrees Cent. were followed over heating periods of 200 hours, in wires drawn from two compositions, (1) cobalt plus 10 per cent tungsten and (2)

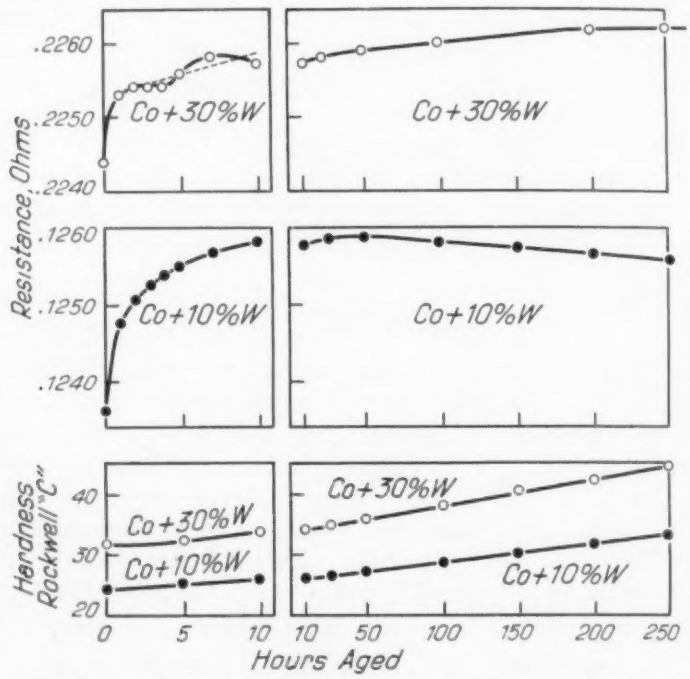

Fig. 1—Electrical Resistance at 25 Degrees Cent. of Cobalt-Tungsten Alloys Quenched at 1050 Degrees Cent. Aged at 500 Degrees Cent.

cobalt plus 30 per cent tungsten. A description of the materials, method of preparation, and method of measurement is given in the foregoing reference.

It will suffice to state here that the wires were first heated for several hours at 1050-1100 degrees Cent. in a hydrogen atmosphere and quickly cooled (also in hydrogen). The resistance of each wire to be aged was measured at 25 ± 0.2 degrees Cent. In groups of three the wires were then heated in a hydrogen atmosphere at 500, 600, and 700 degrees Cent. for a period of one hour, cooled as rapidly as possible, still in a protecting atmosphere of hydrogen and their respective resistances again measured. This operation was repeated at the intervals shown in the curves of Figs. 1, 2, and 3 (this discussion). Three wires of cobalt, from the same metal used in preparing the alloys, were treated and measured as "controls" along with the alloy wires.

The hardness changes corresponding to these treatments are shown in the lower curves of each figure and are drawn largely from data included in the original paper.

Perhaps the most unexpected feature of the resistance curves is the irregular change in the resistance of the higher tungsten alloy during the first ten

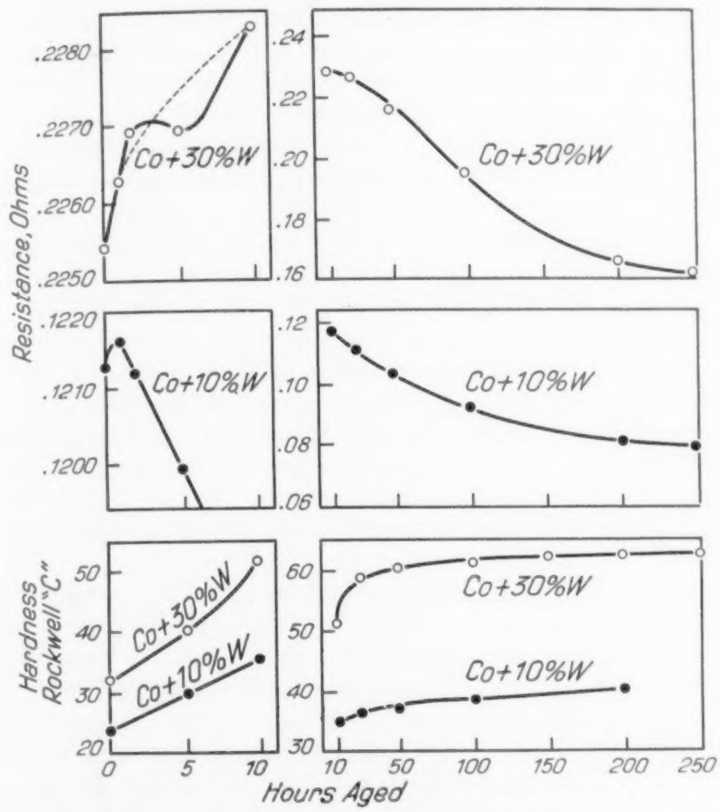

Fig. 2—Electrical Resistance at 25 Degrees Cent. of Cobalt-Tungsten Alloys Quenched at 1050 Degrees. Aged at 600 Degrees Cent.

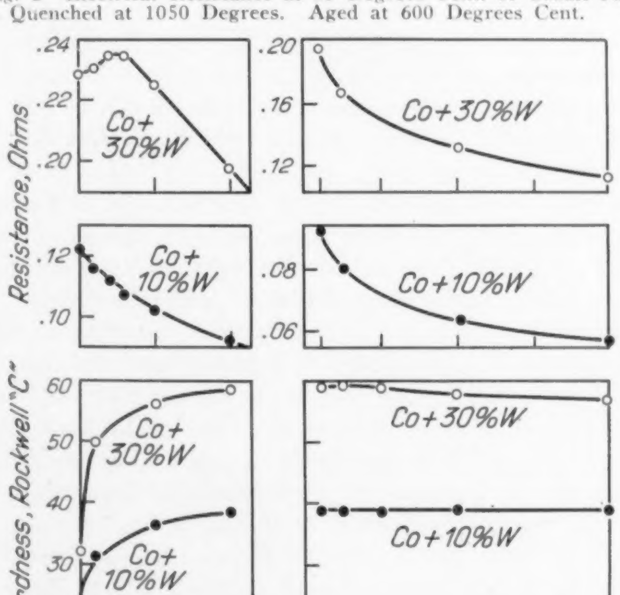

Fig. 3—Electrical Resistance at 25 Degrees Cent. of Cobalt-Tungsten Alloys Aged at 700 Degrees Cent.

hours of aging at 500 and 600 degrees Cent. In these cases two lines have been drawn through the points in question, to indicate that further work is necessary to establish the true shape of the curves. However, it seems proper to state here that no significant changes were observed in the resistances of the cobalt wires heated with the alloy wires throughout the period of aging. This fact eliminates the possibility of impurities as the source of the observed irregularities.

The positive information furnished by the accompanying data may be summarized as follows:

1. The resistance increases substantially during the first few hours at 500, 600, and 700 degrees Cent. in the 30 per cent tungsten alloy. Likewise in the 10 per cent alloy this increase was considerable after the first hour at 500 and 600 degrees Cent. The first measurement of the 10 per cent alloy after one hour at 700 degrees Cent. showed a decrease in resistance, however, suggesting that the maximum in the resistance curve had occurred before the end of the first hour at this temperature.

2. The resistance changes take place more rapidly as the aging temperature is raised.

3. The changes in resistance will be completed more rapidly in the solutions of lowest concentration.

Resistance curves similar to those of the 10 per cent tungsten alloy have been reported by Masing,[2] based on measurements of a beryllium-copper alloy aged at a series of temperatures from 150 to 450 degrees Cent. Masing discusses here the broadening of the X-ray interference lines in the pattern of the solid solution which occurs during the period marked by increasing resistance. This has been taken to indicate the separation of a second phase in a state of atomic dispersion.

It seems likely, therefore, that this initial stage of age-hardening, recognized by its pronounced effect upon the electrical resistance, will be found in the iron-tungsten alloys when these are investigated further.

Written Discussion: By Dr. O. E. Harder, associate director, Battelle Memorial Institute, Columbus, Ohio.

Dr. Harrington's paper is interesting and valuable, not only because of his discussion of age-hardening of alloys, but also because of his discussion of methods of field search as applied to alloy systems.

One type of age-hardening seems to have been omitted from the discussion. I refer to the type of age-hardening as observed in gold-copper alloys. In some respects this type of age-hardening might be included under his classification "C. Complications Due to Allotropy." However, this does not quite cover the case at hand, nor is it clear how the gold-copper alloys could be included in his class "D." In these gold-copper alloys it seems now rather well established that as-quenched or rapidly cooled from the melt or from temperatures above about 400 degrees Cent. the atoms have the face-centered cubic lattice with random distribution of the atoms of the component metals, but upon slow cooling, particularly in the range of 400 to 300 degrees Cent.,

[2]G. Masing, "Heat Treatment of Alloys—Based on Investigations of Light Metals and Alloys of Beryllium." *Transactions* of thirty-sixth general Meeting of the Deutsche Bunsen-Gesellschaft, May, 1931, p. 32.

or upon reheating to within this temperature range a rearrangement of the atoms in the space lattice takes place, resulting in the gold and copper atoms occupying a certain definite position in the lattice, and this rearrangement results in a marked increase in hardness and strength and a decrease in density. This rearrangement is particularly interesting in the alloys of approximately equal atomic concentration of gold and copper atoms (approximately 25 per cent copper by weight). The stable atomic arrangement below about 400 degrees Cent. seems to be face-centered tetragonal with $c/a = 0.93$, or body-centered tetragonal with $c/a = 1.3$. In the face-centered tetragonal arrangement the gold atoms are considered to be at the corners and in the uncompressed faces of the unit cell, while the copper atoms are in the compressed faces. In the body-centered tetragonal arrangement the copper atoms are considered to be at the corners and the gold atoms at the center. These two atomic arrangements are equivalent, as may be seen from examination of models.

This phenomenon is known in certain other binary alloys (for example, copper-platinum and copper-palladium) and may exist in a considerable number of binary alloys. Thus it seems that atomic rearrangement in the space lattice must be added to our already somewhat complicated list of age-hardening mechanisms and thus help to increase the number of alloy systems which must be considered as responding to age-hardening.

The proposed methods of field search for age-hardening alloys are applicable to field search for other types of alloys. Not only are these methods of searching for ternary and more complicated alloys important, but searches for useful alloys in the field of the binary systems are also fruitful. For example, cadmium containing a small amount of copper has been developed in Australia and England as a bearing metal, while in this country cadmium with a small amount of nickel has been advocated by Swartz and Phillips. If it is assumed that we have 40 metals which are available for the production of alloys, then a calculation shows that we have 780 binary systems. As is well known, only a relatively small number of these possible systems have been studied, and doubtless there are many of these in which useful alloys will be found.

Dr. Harrington's discussion of the ternary alloys which respond to age-hardening illustrates how fruitful searches in this field may be. Alloy systems with other desired characteristics may be found by similar searches. As an illustration, cadmium-zinc alloys had certain characteristics which suggested that if the alloys could be modified to have hard particles embedded in a matrix mostly of the cadmium-zinc eutectic the resulting alloy would form a good bearing. The search showed that antimony could be added to these alloys and give the necessary hard particles. Assuming again 40 metals available for combination in ternary alloys, we find that the possible number of systems is 9880. The number of the ternary systems which have been explored is extremely limited. Perhaps not more than about 10 per cent of the possible systems has been given even a preliminary search.

When we come to the still more complicated alloys, such as the quaternary systems, of which the total possible number from 40 metals is 91,390, less than a dozen systems have been studied in anything approaching a complete sur-

vey. These figures are quoted with a view to showing how extensive the unexplored field of alloy systems is and to suggest the importance of using field searches, not only for alloys which show age-hardening but for alloys which show other desirable characteristics.

As a former professor, I regret to note the relatively few alloy systems which are being studied in our educational institutions, where it would appear that the study of these systems, giving scientific or at least basic, fundamental information, might be more in line with the ideals of the educational institution rather than certain applied researches. As is well known, during the past few years a greater portion of our fundamental research on our equilibrium diagrams has come from the industries, good examples of which are the contributions from the General Electric Company and the Aluminum Company of America. Perhaps the student could receive no better training than the experience of working out in a careful and comprehensive way the equilibrium diagrams, and surely the ternary and more complicated systems present a problem which is hard enough for the qualifications for any kind of educational degree. Furthermore, having done a good piece of research work in such a field would seem to be a good recommendation for the candidate to a prospective employer.

Author's Closure

The author greatly appreciates the interesting and pertinent discussions presented by Messrs. Sykes, Harder, and Wood.

Mr. Sykes suggests that types A and B of age-hardening (as classified in the paper) may fundamentally prove to be identical. Professor Wood is more inclusive in suggesting that in the future all methods of hardening will prove to be variations of the same thing. For the present the author does not know of any data giving conclusive proof that atomic solute precipitation (resulting in solvent lattice strain) and critical aggregation of a molecular compound can be treated identically. Both contribute to hardening, but each seems to be distinct in operation. It is always acknowledged that complete understanding of age-hardening must include other tests than those of hardness alone.

In this light it is interesting to consider the electrical resistivity data versus hardening data given by Mr. Sykes. It does not seem illogical to assume that the 30 per cent tungsten-cobalt alloy would be more sluggish of reaction than the 10 per cent tungsten-cobalt. In the period of drawing at both 500 and 600 degrees Cent. from one to ten hours, it is noted that there are two peaks in the resistance curves for the 30 per cent tungsten alloy and but one for the 10 per cent tungsten. Hardness data points corresponding to these resistivity data points are not plotted and may be lacking. As the Rockwell hardness approaches a maximum, however, it is noted that resistivity, as normally expected, decreases. A single peak in the resistivity curve might correspond to an atomic solute precipitation within the solvent lattice while the maximum Rockwell hardness apparently is explained as usual as due to critical grain size of the precipitated hardening phase. Mr. Sykes' data, as he suggests, are similar to those of Masing and quite in agreement with the interpretation.

It is not impossible that absorbed hydrogen (from the preparation of the bars) may be effective in causing the double peak in the resistance curves for the 30 per cent tungsten alloy. The author cannot agree with Mr. Sykes at this time in the general statement that "the cobalt-tungsten system is comparable to the iron-tungsten system" in the same degree, for instance, as the iron-tungsten system may be similar to the iron-molybdenum system. In general, cobalt affects alloy properties quite differently from iron.

Dr. Harder refers to the recently discovered type of age-hardening as exhibited by the gold-copper alloys. As he so clearly describes this reaction the author finds it necesssary only to indicate that this type of age-hardening can be included under number (2) of section "D." This type of transformation is a peculiar one and not as yet well understood. At present it is indicated on equilibrium diagrams in a manner similar to that used to indicate the formation of a compound from a solid solution at a critical temperature above which exists a continuous series of solid solutions toward either side of the compound composition. The lower temperature phase is separated from the higher temperature phase by two two-phase fields that converge at a maximum temperature. It is to be hoped that Dr. Harder's discussion will encourage universities to intensify their student researches concerning equilibrium diagrams.

Professor Wood's general statement is true that cobalt as a simple alloying element (as one of the elements in any binary alloy?) has not seemed as effective as many of the other metals. However, this is apparently not true of alloys of three or more elements; in these, cobalt effects decided changes in the alloy properties.

ON THE MANUFACTURE OF RIMMING STEEL

By William R. Fleming

Abstract

Rimmed steel with which the present paper is concerned, is really a natural steel, nothing being added to it in the ladle to seriously affect the natural physical characteristics of the molten metal. Some typical chemical requirements of different companies are listed herein as are the physical properties which make it particularly desirable for certain uses.

Particular emphasis is placed herein on the physical action of the molten metal in the ingot mold as only by careful study of every heat tapped is it possible to gain definite information between this action and the working of the heat in the furnace. The type of rimming not only tells the story of what has gone before, but predicts accurately what the finished product will be. As yet little is known of high temperature metallurgy and until the arrival of really definite information, the old method of trial and error must be relied upon.

STEEL as manufactured today in large tonnages may be divided into two classes based on the behavior of the molten metal in the ingot mold. If it is effervescent, nervous, restless metal we call it "rimmed steel." If it is dormant and calm, we call it "killed steel." Both are happy and highly descriptive names. One is a natural metal resulting from the simplest process of manufacture; the other is an artificial metal made so by the introduction of agents known to influence the physical condition of both the molten and solidified metal. Here we are concerned only with "rimming steel."

"Rimming steel" has just been referred to as a natural steel; natural because its manufacture involves only a simple melting of ferrous materials, nothing being added in the furnace, ladle or mold that will seriously affect the natural physical characteristics of the molten metal.

In the manufacture of high quality "rimming steel," our first thoughts turn to its chemistry. There is some difference of opinion among manufacturers as to the proper chemistry of "rimming steel"

A paper presented before the Fifteenth Annual Convention of the society held in Detroit, October 2 to 6, 1933. The author is metallurgist with the Andrews Steel Company at Newport, Ky. Manuscript received June 28, 1933.

but in general it is pretty well standardized. The following are typical chemical requirements of different companies for low phosphorus steel:

	Company A	Company B	Company C
Carbon	0.08—0.12	0.06—0.10	0.04—0.08
Manganese	0.30—0.45	0.25—0.40	0.20—0.35
Sulphur	under 0.05	0.05	0.05
Phosphorus	under 0.01	0.01	0.01
Silicon	residual	residual	residual

These typical cases represent distinct opinions held by steel makers as to the chemistry of "rimming steel" and it is possible that each is right under its own peculiar or local condition, since heat size, mold size, furnace practice and other circumstances have a bearing on the proper chemistry of the product. Of this there can be no doubt, namely, there is a vast difference between the rimming qualities and the qualities of the finished product of the two following typical cases:

	Case A	Case B
Carbon	0.12	0.05
Manganese	0.45	0.20
Silicon	residual	residual

These cases are perhaps the extremes of opinions representing general practice. Again it is probable that the great preponderance of "rimming steel" made today falls within the limits of, carbon 0.07 to 0.09 per cent and manganese 0.30 to 0.40 per cent. However, this does not necessarily mean that such chemistry is the final word. It is possible that we are following custom or habit without being sure we are right. The influence of its chemistry upon the physical behavior (rimming) of the metal in the mold and how this physical action bears directly on the quality of the finished product will be discussed later in this paper.

The Physical Character of Rimmed Steel Ingots

Rimmed steel ingots are all more or less unsound, and it is a peculiar type of unsoundness which makes them desirable for specific purposes. The ingot throughout contains blowholes, the number, size and location of which determine definitely the quality of the finished product. Fortunately we can so govern the process of manufacture that the solidified ingot will contain a minimum number of small blowholes located properly. Rimmed steel ingots of the Swiss cheese variety (irregularly located blowholes) are worthless and great

manufacturing losses result from the production of such ingots. The ideal rimmed ingot has a thick or heavy wall or "skin" from one to several inches thick, depending on its size, this shell being physically sound and free from holes apparent to the naked eye. This sound shell (or skin) is the most important requisite for a high quality ingot. Just inside of this thick sound wall are small blowholes regularly distributed and small in number. From these almost to the center of the ingot is an interior sound area and the center again is slightly spongy containing a minimum amount of small blowholes. The number and size of blowholes increases from the bottom to the top of the ingot, but in well made steel not more than 5 to 10 per cent of the ingot need be cropped to eliminate them. In short, it might be said that the physical quality of the rimmed steel ingot depends entirely on the size, number and location of its blowholes. If this is true, it then follows that we should be able to control the manufacturing so as to yield such an ingot. To determine the size, number and location of blowholes, many steel companies have resorted to splitting ingots, a procedure which is very expensive and time-consuming. Much has been learned by this practice, which is post mortem work, by observing in connection with it the physical behavior of the molten mass in the ingot mold so that now the action of this liquid metal gives us a fairly accurate picture of the inside of the solidified ingot. Thus we are able to observe carefully a heat in the molds and pass judgment right then as to its fitness for the product intended.

Behavior of Open or Rimming Steel in Molds During Process of Freezing

The action of metal in the molds is varied and each different type of action represents a distinct quality of ingot, that is, the action which we observe tells us the secret hiding places of the blowholes within the ingot. The ideal ingot with thick skin and deep-seated blowholes will be considered first.

Example No. 1—Assume an ingot mold 20 x 22 weighing about 7000 pounds. The metal is poured through a 2-inch nozzle and the mold filled in about fifty seconds. The mold is filled within three inches of the top and the molten mass remains exactly where it finished pouring—it neither sinks nor rises. Within less than one minute the metal begins to freeze along the shore line; it starts rimming. It continues rimming straight across until it establishes a flat

shore line one to several inches wide. From this point the metal gradually sinks, forming a declining bank to the center of the ingot. Rimming continues until freezing is completed, leaving a slight depression in the exact center. Having observed such action of any given heat in the molds we can confidently predict that the ingots will be thick-skinned, the number and size of blowholes will be small and they will be deep-seated. Furthermore, the sheets rolled from such ingots will be practically free from seams, scabs, and blisters, provided of course, the steel is not subsequently mistreated in the steel plant or sheet mills.

Example No. 2—Conditions are like Example No. 1 up to the final setting of the ingot. In this case a well about three to five inches in diameter in the center stops rimming, the steel becomes mushy, builds up in the center, forming a projection ("nigger head") which, if not capped, sometimes grows in height as much as twelve inches or more.

Having observed such ingots we can predict: Thick-skinned ingots with increased number and size of blowholes in the upper central portion. Sheets from such ingots will be free from seams and scabs but will contain a greater number of blisters than Example No. 1. A larger crop from the top of the ingot would be necessary, but fortunately a close inspection of the sheet bars reveals the amount of discard demanded. It is perhaps true that a majority of rimming steel heats behave in the molds as described here under Example No. 2. The practice generally is to prevent this final building up of the center by capping, which is done to facilitate stripping. However, liberation of gases still continues within the ingot and not being able to escape, they form larger blowholes in the top of the ingot.

Example No. 3—In this case the metal starts rising in the mold shortly after teeming and continues to rise gradually while rimming, until a maximum of from a fraction to several inches has been reached. In most such cases the ingot rims straight across. Such conduct in the molds indicates thin-skinned ingots and excess blowholes located at random throughout the ingot. Ingots from such a heat yield an excessive number of seamy and blistered sheets and unless the heat can be diverted into a cheap grade of sheets, it would probably save considerable money and grief to remelt it. Assuming that such an ingot stopped pouring at 66 inches and grew to 70 inches while rimming, it is clear that its volume increased greatly. The approximate volume of the liquid at 66 inches was 29,040 cubic inches

while at 70 inches the volume was 30,800 cubic inches, an increase of 6.6 per cent or almost one cubic foot, which means that the interior of the ingot contains one cubic foot of blowholes in excess of its normal quota. Occasionally the production of such heats is excusable but in the main they can be avoided.

Example No. 4—Assume in this case that the metal starts rimming at the exact height at which teeming stopped and continues to rim straight across, leaving a flat top, with no tendency to nigger head in the center. Such ingots are also of high quality, having a thick skin and a minimum of small blowholes. They differ from ingots in Example No. 1 in that they contain more widely distributed blowholes through the interior of the ingot. Sheets rolled from such ingots will inspect well although the number of seams and blisters will be slightly higher than in Example No. 1.

The types of rimming steel ingots just described cover the general characteristic differences in heats encountered in daily steel mill practice. There are of course freak heats whose antics in the molds do not come under any code of ethics and such heats will not be discussed in this paper.

Let us now pass to the consideration of the steel-making process itself in an effort to determine the cause of various behaviors of different heats in the molds, which in turn means heats of various degrees of quality.

The chemistry of rimmed steel has already been discussed to some extent. This chemistry has a direct bearing on the action of liquid metal in the molds but in most cases it is not solely responsible. Other conditions jointly contribute, but before discussing these jointly, let us first consider the influence of chemistry alone.

It is well known that the carbon and manganese must be low in rimming steel. The maximum of these two elements for good rimming action in the molds is close to 0.16 per cent carbon and 0.50 per cent manganese. In fact it is doubtful if such steel is ever a true rimming steel. It does of course rim after a fashion, indeed some have claimed to rim steel as high as 0.30 per cent carbon, but the character of rimming exhibited by such steel is not the type of rimming dealt with in this paper; it is not true rimming. For our present purpose, true rimming steel may be considered as described under Example No. 1; that is, liquid steel which neither rises nor sinks from the final pouring level; which forms a flat shore line one or more inches wide; which then begins sinking, forming about a

45-degree slope to the center of the ingot; and which at its final freezing, forms a slight cavity at the exact center. Such ideal rimming, according to the writer's experience, demands very low carbon and manganese. It must be below 0.08 per cent carbon and 0.30 per cent manganese, and to accomplish this ideal rimming consistently heat after heat, the carbon and manganese should be below 0.05 and 0.15 per cent respectively. As the carbon and manganese increase the tendency of the metal to rise increases, the character of rimming changes and the final ingot gradually decreases in quality with reference to advantageous location of blowholes. It, of course, is recognized that in many cases the presence of set minimums of carbon and manganese is necessary and furnace practice must be varied to accomplish the best rimming under these conditions. However, it is not possible to consistently rim steel in the higher carbon and manganese range and produce ingots of quality comparable to those of lower range.

It is not the purpose of this paper to discuss the chemistry and mechanics of rimming in the mold; in fact little is known about it, except that evolution of gases takes place during the process of freezing, which explains partially the mechanics of this phenomenon. The exact chemistry producing these gases is not known.

Other elements than carbon and manganese influence the chemistry and mechanics of rimming. Some aid spontaneous evolution of gases and thereby improve the quality of ingots, while others hinder such evolution to the detriment of ingots.

It is possible that oxygen is largely responsible for the phenomenon known as rimming. This is in harmony with the fact that increasing carbon and manganese decreases rimming action, not because of direct influence of those elements but because their presence prevents the metal absorbing oxygen. The amount of oxygen held in solution in liquid steel of very low carbon and manganese content is probably very small. If these elements are reduced by excessive oring to 0.01 per cent, the solidified ingot will be found to contain less than 0.10 per cent of oxygen by weight. A rimming steel of 0.05 per cent carbon and 0.15 per cent manganese usually contains less than 0.01 per cent of oxygen, the quantity evolved as carbon monoxide during the rimming period being unknown.

Aluminum is perhaps the most influential element added to rimming steel, and its proper use by the skilled steel maker many times converts an inferior or mediocre heat into one of high quality.

No two heats require the same amount of aluminum, a few pounds more or less in the ladle often being responsible for the steel being good or bad. Aluminum has a powerful influence when added to molten steel in small quantities. The exact nature of the chemistry of this action is unknown, although it is common knowledge that it has a great affinity for oxygen, at high temperature. Its action in molten steel is commonly supposed to be deoxidizing but this by no means explains the phenomenon that often is observed when exceedingly small quantities are used. If liquid steel in an ingot mold happens to be impregnated with finely divided gases, an ounce or two of aluminum will cause it to sink several inches almost instantly. This phenomenon cannot be attributed to deoxidation. It suddenly in some mysterious way greatly increases the solubility of the entrapped gases in the steel and it is this influence upon the solubility of gases which makes aluminum almost indispensable in the making of rimming steel. Aluminum should always be added in the ladle and always sparingly. Too much results in sluggish rimming which in turn means thin-skinned ingots yielding seamy and blistered sheet, while if too little has been added in the ladle it can be used correctively in the molds. This use of aluminum in the ladle has a very definite relation to the working of the heat in the furnace and the skillful and experienced melter can instinctively vary his aluminum additions to suit the needs of each particular heat. To acquire such skill the melter must study the action of each heat in the molds and if such action is abnormal he should trace it directly from the working of the heat in the furnace and its condition when tapped. A properly worked heat finishing about 0.05 per cent carbon and 0.20 per cent manganese should never require the addition of over 20 pounds of aluminum per 100 tons.

Silicon possesses to some extent the properties of aluminum when used in small quantities. However, its use in rimming steel to take the place of aluminum is not advisable.

Titanium is used to some extent in rimming steel. When used in proper quantities it leads to improved rimming and consequently improved ingots.

Phosphorus when added in the form of ferrophosphorus in the ladle affects the rimming qualities adversely. The steel has a marked tendency to rise in the molds, and rimming is less defined and inclined to sluggishness.

Manganese when added in the ladle, like phosphorus, affects

rimming adversely. Since either phosphorus or manganese must be added to most open steel, the condition of the bath at tapping must be carefully regulated to counteract, as far as possible, the handicap to proper rimming resulting from the additions.

Furnace Practice

Open-hearth furnace practice followed in the production of rimming steel is, in general, pretty much the same in all steel plants. It can be made from various mixtures of pig iron and scrap or from scrap alone, but this is governed by local conditions and is outside the scope of this paper. Acid open-hearth practice will also be omitted since the tonnage of such steel rimmed is exceedingly small.

In most steel plants the production of rimming steel is a matter of routine. Tonnage is the main consideration although quality receives more or less attention in some plants. Of course, in the production of special sheets a certain amount of attention must be given to open-hearth practice since this is chiefly responsible for rejections due to steel defects. But in spite of this the writer is convinced that great sums of money are lost yearly by most steel plants through lack of careful and proper open-hearth supervision. This does not imply that the open-hearth itself is at fault. On the contrary, company policy is usually responsible by demanding low costs and large production from the open-hearth. In turn company policy is often dictated by cut throat competition which has been so devastating to the steel industry in the past few years. Yet no one can afford to blindly fail to recognize the fact that there exists, between low costs and high production on one side and quality on the other, a well defined dividing line on either side of which lies economical waste. The old homely saying, "You can't make a silk purse out of a sow's ear," is not literally true. A certain definite percentage of sow's ears may be mixed with silk "scrap" to produce a silk purse of a high standard of utility and beauty. That is the steel-maker's problem—what percentage of sow's ears can be used.

While the open-hearth furnace is essentially a melting furnace, it is also a refining and finishing furnace. If used only as a melting medium, tonnage records are easy to break, but, if proper attention is given to refining and finishing, tonnage will lag slightly behind. The latter is sure to pay higher dividends.

The working of a heat of steel in an open-hearth furnace, after the charge is melted, may be divided into three periods: oring, refining, and finishing.

Oring generally is not given the serious consideration it deserves. It is something more than a dumping process and waiting until it "has gone." Ore requires a comparatively short time until it has apparently been consumed. If fed on this principle, most rimming steel heats will be over-ored at the finish which introduces variables almost surely leading to grief in the finished product. Each feed of ore must be given its proper time which is governed by slag and temperature conditions mainly. The slag must not be allowed to grow too thin while oring and the temperature must be kept properly adjusted to the high side.

After the oring has been completed the viscosity of the slag must be increased to its proper consistency as rapidly as possible by the addition of burnt lime. A moderately heavy slag is necessary.

During the slag finishing period close attention must be given to the temperature of the metal, which should be tapped on the hot side. If the melter has paid careful attention to oring, slag building, and temperature, he is prepared to use aluminum intelligently in the ladle, and the resulting ingots will rim properly and yield a satisfactory product.

The manufacture of rimming steel today is an art, not a science. Little is known of high temperature metallurgy and until the arrival of more definite information concerning the chemical reactions and relations of metal, furnace bottom, slag and temperature, we must rely on the old method of trial and error. In this paper emphasis has been placed on the physical action of molten metal in the ingot mold. It is here that the wide awake and alert melter must find the answer to his efforts put forth on the heat while in the furnace and ladle. This character or type of rimming not only tells the story of what has gone before, but it predicts accurately what the finished product will be; so that the real steel maker is the man who makes a careful study of the action in the molds of every heat he taps and is able to form a definite opinion between this action and the working of the heat in the furnace. The man who simply taps his heat and relieves himself of responsibility as soon as the ladle is taken away is a melter, not a steel-maker.

DISCUSSION

Written Discussion: By H. H. Ashdown, metallurgist, Westinghouse Electric & Manufacturing Co., East Pittsburgh, Pa.

I have been invited to discuss this paper. The preparation of any paper entails a great amount of work and every author is deserving of credit and encouragement. The author no doubt has recorded the results of his observations and experience but to me in this presentation there appears to be little information of added interest. It would appear that the author is neither conversant with the work of Herty or the publication of the British Iron and Steel Institute on the heterogeneity of steel ingots incorporating much European practice. Reports No. 4 and No. 5 in themselves contribute valuable research and investigation work on rimming steels and, in addition, the appendix has an excellent bibliography of the authors of papers on this subject.

Due to the greatly increased application of rimming steel during recent years the location and type of gas holes in the ingots have become of marked importance.

It is well known that these holes are largely the result of the dissolution and reaction of oxygen during solidification and that the extent of their presence is due largely to the finishing temperature in the furnace controlling the balance of FeO in the bath and slag. These two fundamentals control the final product although this to some extent is also influenced by the speed of pouring and consequent rate of solidification.

The use and extent of a deoxidizer such as aluminum needs little comment on its effect on the quantity and to some extent the disposition of the gas holes in the ingots. Size and weight of the ingots have a marked effect on the quantity and location of the gas holes, the larger ingots due to their slower cooling retain greater quantity and thus require more deoxidation.

Although the author on page 534 states that a 2-inch nozzle is used for casting 20x22-inch ingots weighing about 7000 pounds which are presumably intended for sheet work, the paper is not specific as it also mentions large tonnage production and may be considered in a general sense. It would be of interest to learn the size of nozzle used for large ingots intended for plate work, as much of this material is now used for heavy fabrication work wherein excessively oxidized internal surfaces and mass inclusions adversely affect fusion welding.

To me, on page 540, the author's closing remarks appear to be particularly unfortunate. I do not think it will be generally accepted that the manufacture of rimming steel today is an art, not a science and that this product still is the result of trial and error. Such a statement does not give the consumer much confidence in the product he is fabricating.

The publications already referred to show in many instances this material is in a large measure under scientific control and is in the hands of competent metallurgists.

It also is my opinion that the final stage of steel making should be in the ladle and not in the mold where due to indiscriminate addition of aluminum large quantities of aluminum oxide (Al_2O_3) are enclosed in the steel often with unfortunate results to the finished product.

It may be of interest to record that of quite recent date, I have seen

several instances where this material has been machined and when put under an air pressure test of 60 pounds per square inch, it has leaked through thicknesses varying from ½ inch to 2 inches both longitudinally and transversely.

On a deep etch these materials have disclosed not chain lines, but plates of aluminum oxide enclosed in the cavities and for air to freely percolate through considerable thicknesses of this material indicates that much is to be desired in its manufacture to meet its present day application.

Written Discussion: By H. L. Geiger, chief metallurgist, International Harvester Co., South Chicago.

I was interested to read this paper by Mr. Fleming which I am sure will be of interest to steel users as well as steel producers, particularly those associated with the sheet industry. Mr. Fleming classifies rimming steels, first, according to chemical analysis and cites two extreme types which he designates as Case A and Case B. It is the writer's conclusion from his observation of these two extreme types of steels, that their behavior is entirely different in rimming, hot working and cold working. In Case A (C 0.12 per cent — Mn 0.45 per cent), the steel as a rule requires no aluminum in the ladle and very

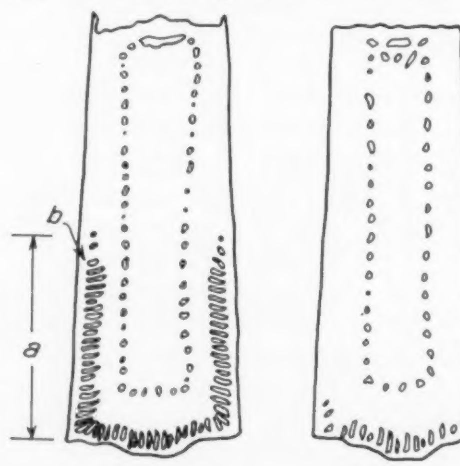

Fig. 1

little in the mold. Some heats of this type which require two to four ounces of aluminum per ton in the mold in the first part of the teeming operation will finish pouring using no aluminum to give the proper rimming action. As a rule in the higher carbon (0.14 per cent max.) and higher manganese (0.50 per cent max.) range the steel has a tendency to rise in rimming as under Mr. Fleming's Type 3. As the manganese and carbon are lowered there is less tendency to rising. In the range of 0.30 to 0.50 per cent manganese, the depth of skin is not so much determined by the manganese content as it is by the temperature at which the heat is tapped and teemed. The author states that the heats must be tapped on the hot side. In the writer's experience with the 0.30 to 0.50 per cent manganese range, the ingots with the thickest skins are obtained with the "cooler" pouring temperatures and incidentally have fewer seams. By "cooler" is not meant "cold" heats which leave very heavy skulls however.

Considering now Case B (C 0.05 per cent, Mn 0.20 per cent): This type of steel has a milder, more moderate rimming action, usually not rising or falling, but holding to the height at which it finished pouring. In this type of heat it is probably safe to add aluminum to the ladle to the extent of ¼ pound per ton, although it is not necessary. The writer cannot agree with Mr. Fleming regarding ladle addition of aluminum. It is far better to judge the amount of aluminum needed by the action of the steel in the first mold or two, and govern the additions to the molds accordingly. Too often what would have been a good rimming heat has been ruined by ladle addition of aluminum, resulting in a sluggish heat and often a seamy heat. Mr. Fleming suggests that the furnace-man can judge the amount of aluminum to add to the ladle by the action of his heat in the furnace. This is sometimes difficult to do, as rim steel making is too much of an art and not enough of a science for the steel maker to judge correctly in most cases. Some effort has been made to control slag FeO as a criterion of what should be expected in rimming, but this is not reliable as there are other factors contributing to rimming. Also the slag FeO is not always a criterion of the FeO in the metal. Hence, efforts to put rim steel making on a scientific basis are not fully successful as yet.

Next, considering the types of ingots obtained from these two extreme cases: Fig. 1 presents the type of structure which is to be expected from heats of A and of B poured at about the same temperature. Considering structure of A: As the temperature of teeming is lowered, the height of the outer shell of blow-holes (a) is lowered, while the thickness of the skin (b) increases. In the case of the B analysis, temperature is not so important, as the outer layer of blowholes does not have a tendency to form and, as a result, this type of steel may be poured over a greater range of temperature with the same results. Other characteristics of these two steels may be tabulated as follows:

	A	*B*
Ingot Rolling	Best 1950-2100 degrees Fahr.	Best over 2100 degrees Fahr.
Behavior in Rolling	Tendency towards seams but general surface is smooth.	Absence of seams but is red-short leaving a rough surface.
Sheets	Smooth surface, but tendency towards seams and when pickled shows tendency towards pickle blisters.	Uneven thickness due to red-shortness but absence of seams. Shows no tendency towards blistering in pickling.
Cold Forming	Good; fairly ductile, depending on anneal.	Good; very ductile, showing no signs of cold-shortness.

Mr. Fleming is probably correct when he states that little is known of molten metal reactions at high temperature. We may find an example to illustrate this difference in the "rodding" of a heat. When a relatively cold steel rod is put into a molten bath a violent boil takes place in the vicinity of the rod as it travels back and forth in the molten metal. Of what takes place in the metal at this point, we are certain of only one factor, namely that the molten metal is cooled down in the vicinity of the rod. This cooling increases the boil, liberating great volumes of gas. The question in the mind of the metallurgist is (a) does the lower temperature increase the rate of the following reaction:

$$\text{(1)} \qquad FeO + C \rightarrow Fe + CO \text{ (gas)}$$

thus increasing the rate of CO given off; or (b) is the temperature of the metal in the vicinity of the rod lowered to a point below the critical solubility of CO gas liberating this gas dissolved in the molten metal as CO? Since the solubility of gases in most molten metal increases with an increase in temperature, all logic points to (b) as being the case.

To make it more clear, when a steel bath is burning out carbon and giving off CO gas only part of this gas escapes in the bubbling action, while part may be retained in solution in the molten metal. The amount in solution will represent the saturation point of that gas in the metal for the existing bath temperature. Thus, the gas which is being given off from the bath by reaction (1) would be in constant excess of the amount needed to maintain the constant saturation of the metal.

If the foregoing idea or thought on gas evolution is transferred to the molds where effervescing continues as in the furnace, we find that the more violent reaction evolving gases takes place along the freezing surface. Here again we have a chilling surface (as in the case of the rod), producing the greater evolution of gas and pointing once more to the solubility of CO gas in the liquid metal as being mainly responsible for the rimming action.

However, before drawing conclusions too hastily from this example, it might be well to approach this problem from another angle. It is generally known that an ordinary ingot during rimming gives off 25 to 35 times its own volume in gas which consists mainly of CO with some small amounts of hydrogen and nitrogen. Of these three gases the solubility of hydrogen in molten iron is known, being only 2.10 times the volume of the metal. It is very likely that the CO and nitrogen are not much more soluble than hydrogen, since hydrogen is much more soluble in the solid steel than either of the other gases. This points to the possibility of only a small portion of the evolved gas in rimming as being in solution as a gas and the large portion of the gas evolved would then be a product of reaction.

Further study of the solid steel points to reaction (1) as being the predominating reaction in the mold.

Fig. 2 is a deep etched billet rolled from an ingot which was poured at a medium temperature of C 0.07 per cent and Mn 0.37 per cent analysis. This heat had an FeO content of 0.220 (0.048 per cent oxygen) in the bath. Average analyses of the cases made for oxygen of these heats show 0.015 to 0.020 per cent while the core shows a presence of about 0.040 to 0.050 per cent oxygen. Analysis of the case for carbon reveals 0.04 to 0.05 per cent carbon, while the core will reveal about 0.07 per cent. An average for the ingot in carbon will be about 0.055 per cent, while an average for oxygen will probably be about 0.030 per cent. From these results it is evident that there is a reaction which advances along the wall of the freezing metal which removes oxygen and carbon, probably reaction (1). As the temperature of the remaining liquid metal (represented by the core in Fig. 2) reaches the mushy stage, it solidifies as a mass, retaining in solution the remaining oxygen which is nearly equal in amount to the oxygen in the metal as poured. The dark core etch is probably due to the dissolved oxygen entrapped in the suddenly frozen metal.

Thus we see where the first idea favors the solubility of gases as being responsible for a large part of the rimming action, while the latter idea de-

rived from a study of the solid metal points favorably to reaction (1) as being largely responsible for the rimming action and therefore we must agree with Mr. Fleming that the reactions in the molten metal are not fully understood.

Before closing my discussion I want to say a word regarding the author's statement, "A rimming steel of 0.05 per cent carbon and 0.15 per cent manganese usually contains less than 0.01 per cent oxygen." The writer has found that, as a rule, the oxygen content of metal is inverse to the carbon of the steel

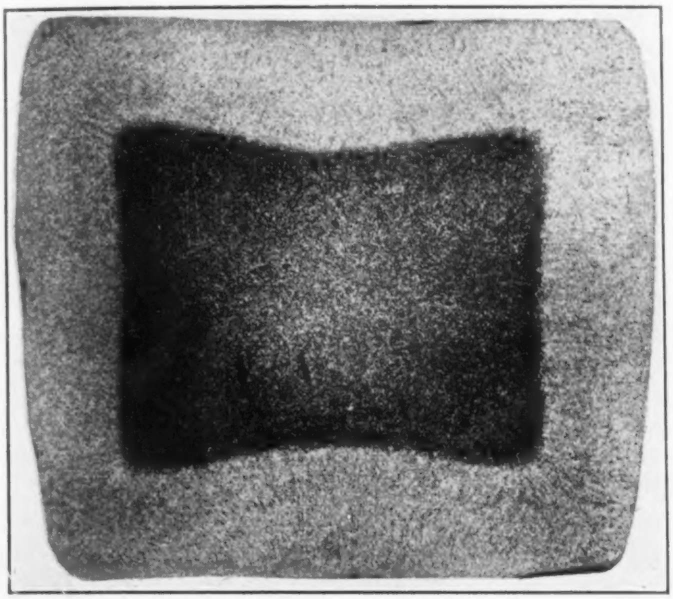

Fig. 2—Light Outer Area is Referred to as Rim, Dark Center is Core.

prior to tapping and a low carbon heat is usually pretty high in oxygen (0.05-0.08 per cent). It is an erroneous though natural belief among some steel makers that the lower carbon rimming steel (0.02-0.05 per cent) is low in oxygen, based entirely on the less vigorous rimming action obtained on these steels as compared with the more vigorous rising steel of 0.09-0.12 per cent carbon range. The difference, however, is not in the oxygen, but in the carbon content. In the low carbon steel (0.020 per cent carbon) there is not enough of a carbon concentration to produce a vigorous action. The writer knows of a steel plant some years ago, making soft rimmed steel of 0.02 to 0.04 per cent carbon for wires. Because of the lack of action in the rimming, fine ore was put in each mold in an effort to furnish "enough" oxygen to produce a rimming action. The deficiency of carbon was overlooked, with disastrous results as far as the rimming and quality of the steel were concerned.

Author's Closure

I have read with much interest Mr. Geiger's able discussion of my paper. My only regret is that the paper did not provoke many other discussions of similar high calibre by other experienced steel makers.

Our knowledge of steel making is built up more rapidly by free exchange of ideas among both metallurgists and practical steel makers, and Mr. Geiger is to be congratulated on his frank and able discussion of this paper.

Mr. Gieger states that rimmed steel should be tapped on the cold side. Since he does not add aluminum in the ladle, a cooler temperature is probably preferable, but steel tapped on the hot side, treated with aluminum in the ladle, most certainly yields best results in our plant. A small skull (very thin pancake not covering entire bottom) is not objectionable but no skull is preferable. In our practice at least 10 pounds of aluminum per 100 tons is always safe to use in the ladle with a 0.12 per cent carbon, 0.40 per cent manganese heat.

The ingot shown as case "A" would be considered below par in our plant for a 0.12 per cent carbon, 0.40 per cent manganese heat. Blow holes are too large and located too near the surface. The lower half of such an ingot would be too seamy and very likely produce excessive blisters in pickling. This sketch depicts my chiet objection to using aluminum in the molds. Steel properly tapped and treated with aluminum (or titanium) in the ladle will not have such large, numerous and shallow seated gas holes.

Mr. Geiger's statement "as the temperature of teeming is lowered, the height of the outer shell of blow holes (a) is lowered, while the thickness of the skin (b) increases," may be true with his practice but it most certainly does not hold in all practice—or is it possible that the omission of aluminum in the ladle would be the answer? Mr. Geiger is correct in stating that too much aluminum in the ladle is fatal to good rimming qualities, but it is an easy matter for a good alert melter to keep his aluminum on the safe side.

Mr. Geiger states that the oxygen content of the lower carbon rimmed steels is usually 0.05 to 0.08 per cent. If by "lower carbon" he means 0.02 per cent carbon and 0.02 per cent manganese, then his figures are approximately correct, but if he is referring to 0.04 to 0.06 per cent carbon and 0.15 to 0.20 per cent manganese then the percentages are very much too high. It is not difficult to make 0.05 per cent carbon and 0.15 per cent manganese steel under 0.02 per cent oxygen and it is not difficult to make 0.02 per cent carbon and 0.02 per cent manganese steel under 0.05 per cent oxygen. These figures, of course, refer to drillings taken from a rolled section of an ingot and not to a test taken from the bath, prior to tapping. An accurate sample for oxygen determination cannot be obtained in this manner.

Mr. Geiger cites a particular heat of 0.07 per cent carbon and 0.37 per cent manganese which contained 0.048 per cent oxygen in the bath. To me such a condition is unthinkable as I have never known of a heat of similar analysis to exceed 0.02 per cent oxygen.

THE LIFE OF TURNING TOOLS AS INFLUENCED BY SHAPE

By O. W. Boston and W. W. Gilbert

Abstract

The relation between the cutting speed and tool life for a given tool when cutting a given steel under constant conditions has long been known to follow the equation $VT^n = C$. *This relation has been shown at different times to hold for tools of different compositions when cutting a steel, and tools of a constant type when cutting various steels. There is a wide range in value of exponent and constant, however, as factors such as tool material and shape, material being cut, size and shape of cut, etc., are varied. The relative importance of these factors has not been clearly defined.*

This paper gives the results of an investigation to determine the value of the exponents and constants of the tool-life cutting-speed equation, $VT^n = C$, *for a given tool of high speed steel when cutting an S.A.E. 2335 steel forging. All tests were run dry and the data submitted are from cuts 0.100 inch in depth and 0.0125 inch feed per revolution.*

The variables studied were the tool-nose radius, the side-cutting angle, the side rake angle, and the back rake angle. It is shown that there is a definite relation between each of these variables and the exponent and constant of the equation.

Introduction

THERE has been a great deal of confusion as to the relation between the cutting speed and tool life per grind in turning. Taylor's classical work(5)* concludes that cutting speed $\times$ tool-life$^{1/8}$ = a constant, or $VT^{1/8} = C$ (1)

The Bureau of Standards' experiments(2) by French and Digges

*The numbers in parentheses refer to the corresponding numbers in the selected bibliography appended to this paper.

A paper presented before the Fifteenth Annual Convention of the society held in Detroit, October 2 to 6, 1933. Of the authors, O. W. Boston, a member of the society, is director, Department of Engineering Shops at the University of Michigan, Ann Arbor, Mich., and W. W. Gilbert is research fellow of the graduate school of that university. Manuscript received June 26, 1933.

conclude that the exponent should be 1/7. Taylor's results were obtained when cutting a carbon steel with a tensile strength of 70,000 pounds per square inch, with his so-called round-nose high speed tools of the forged type, the width of the shank of which was ⅞ inch. The radius of the nose of Taylor's tool is found to be $\frac{3}{32}$ inch from his formula

$$R = \frac{A}{2} - \frac{5}{32} \tag{II}$$

in which R is the radius of the nose and A is the width of the shank, both in inches.

The tests of the Bureau of Standards were based on forged tools of high speed steel, the section of the shank being ¼ inch wide by ½ inch deep, with a radius of ⅛ inch. A 3.50 per cent nickel steel, having a tensile strength of 100,000 pounds per square inch, was cut. The tools of both experiments had 6 degrees clearance, 8 degrees back rake, and 14 degrees side rake. Taylor used a depth of cut of $\frac{3}{16}$ inch and a feed of $\frac{1}{16}$ inch, while the Bureau of Standards used $\frac{3}{16}$ inch depth of cut and 0.028 inch feed.

A different form of tool was used by the Bureau of Standards(3) when taking finishing cuts. Exponents between $\frac{1}{10}$ and 1/13 were obtained when the cuts were 0.010 inch deep and the feed was 0.0115 inch per revolution of the work. The trailer tool method was used for determining the time of tool failure in these finishing cuts which may account for the difference in the tool life cutting speed relation. Whether the change in exponents in the Bureau of Standards' experiments for rough turning and finish turning was due to the different method for determining tool failure, different shape of tool used, or to the difference in the size of cuts has not been clearly shown. Also, the Bureau of Standards' tests have shown(1) that, when cemented tungsten carbide tools were used for taking roughing cuts, the exponent was 1/5. Again a tool of different form was used. The back rake was 0 degrees, side rake 14 degrees, end-cutting angle 10 degrees, clearance 6 degrees, and the nose radius $\frac{1}{16}$ inch. A 1.75-inch radius cutting edge was used. The depth of cut was 0.1875 inch and the feed was 0.031 inch. The difference in exponents might be due to the tool material or to the tool shape.

Judkins and Uescher(4) show that, when using Firthite type

T64 tools of cemented carbide containing tantalum carbide, having a side rake of 12 degrees, a back rake of 6 degrees, front and side clearance of 5 degrees, and $\frac{3}{32}$-inch nose radius, having a curved cutting edge of the conventional roughing type, the following values of n were obtained:

When cutting S.A.E. 1040 steel with $\frac{1}{16}$ inch depth of cut and 0.025 inch feed per revolution, $n = 1/6.22$

When cutting S.A.E. 1060, $n = 1/6$

When cutting 3.50 per cent nickel steel, $n = 1/6.2$

When cutting high carbon high chromium stainless steel, $n = 1/6.4$.

Many other illustrations similar to these can be given to show that the value of the exponent of tool-life in the equation giving a relation between cutting speed and tool-life differs from 1/7 to ⅛. There has been so much discussion in connection with the correct value of n when cutting steel with high speed steel tools that the authors made this the subject of a special investigation.

Object of the Investigation

The object of this investigation was to determine for a given tool of high speed steel, when cutting a given steel forging, the value of the exponent of the tool-life cutting speed equation. The experiments were performed with a constant depth of cut and feed. After determining the exponent and equation for the given tool ground to one shape, the equation was redetermined for the tool ground to a number of other shapes, keeping the tool material, the material cut, the feed, and depth of cut constant. In this way, the influence of the change in tool shape could be determined on the cutting speed tool-life equation $VT^n = C$, as shown by the corresponding values of n and C. The effect of tool angles on tool life also was studied.

The Material Cut

The material cut was purchased as a forging of steel corresponding to the analysis of S.A.E. 2335. The bar was 48 inches long and 13 inches in diameter, weighing some 2200 pounds. The chemical analysis was as follows:

Carbon	0.36	Phosphorus	0.032	Silicon	0.10
Manganese	0.50	Sulphur	0.036	Nickel	3.29

After forging, it was air cooled from 1460 degrees Fahr., and furnace cooled from 1240 degrees Fahr. It had a yield point of 65,600 pounds per square inch, a tensile strength of 101,250 pounds per square inch, elongation 25.5 per cent in 2 inches, reduction of area 53 per cent, and a Brinell hardness of 207. The Scleroscope hardness, as determined on its periphery at various points from end to end, averaged 28 to 30.

The Cutting Tools

For the tests covered by this report, 100 tool bits were obtained of the regular 18-4-1 type of high speed steel. These bits were ⅜ inch square and 3½ inches long when new. They were carefully selected from the same heat, and heat treated under special conditions, so that they could be duplicated as desired. The analysis of the steel is as follows:

Carbon	0.76	Phosphorus and Sulphur	0.018
Manganese	0.34	Vanadium	1.08
Silicon	0.26	Tungsten	18.22
Chromium	4.06		

The bits were preheated to 1600 degrees Fahr. and then raised to a high heat of 2400 degrees Fahr. until sweat appeared. They were drawn two hours at 1100 degrees Fahr. in lead. All bits were tested for uniformity of hardness and found to average from 64 to 67 Rockwell C.

The tools were carefully ground on a universal tool grinder under a copious stream of an emulsion. All burrs were removed by light hand honing after grinding. The tools were used in sequence, being reground and used again so as to reduce the variation of tool material to a minimum. A special tool holder was used for holding the tool bits in a lathe. The tool bit and tool holder are illustrated in Fig. 1. The names of the various angles are indicated. This particular tool, shown as No. 7, Table I, is referred to as 8-14-6-30-3/64 R. These values in the order named are, back-rake angle 8 degrees, side-rake angle 14 degrees, front- and side-clearance angles 6 degrees, side-cutting angle 30 degrees, and nose radius $\frac{3}{64}$ inch. Table I lists the various angles and nose radii used on the fifteen different tool shapes tested. In grinding the nose radius,

the tools were first ground to a point, after which the radius was ground by hand and checked with the radius gage. This method was found to give consistent results.

The Tool-Life Tests

Values of tool life T in minutes were obtained experimentally for corresponding cutting speeds V in feet per minute when cutting

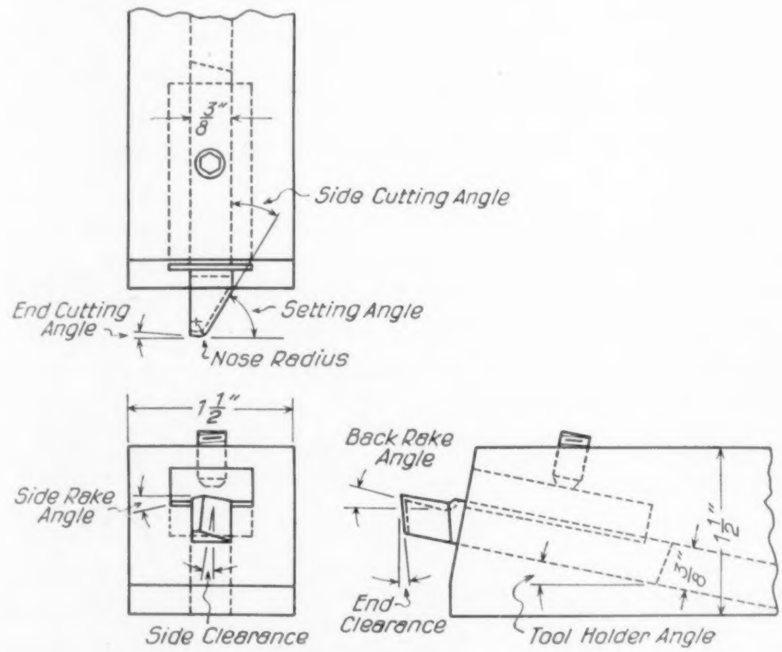

Fig. 1—Nomenclature of Tool Bit and Holder.

the S.A.E. 2335 steel forging with the regular 18-4-1 high speed steel bit, ⅜-inch square, as ground to shapes shown in Table I. In all tests, the depth of cut was 0.100 inch and the feed per revolution of the work was 0.0125 inch.

For each tool shape, the values of V and T for various cutting speeds were plotted on log-log paper. It was found that for each case a straight line resulted, showing that the logarithm of V varies inversely as n times the logarithm of T. This gives rise to the equation

$$VT^n = C \qquad \text{(III)}$$

in which V is the peripheral cutting speed of the work in feet per minute on the surface of the work, T is the tool life in minutes

Table I

A List of the Tools Used in the Tests, Together with the Values of n and C as Determined from the Cutting-Speed Tool-Life Equation $VT^n = C$. A Key Developed to Indicate Briefly a Given Tool is Indicated. For Example, Tool Number 2 is Designated as 8-14-6-0-$\frac{1}{32}$-R.

Tool Number	Back Rake Angle	Side Rake Angle	Side and End Clearances	Side Cutting Angle	Nose Radius in Inches	n	C
1	8	14	6	0	0	1/13.1	101
2	8	14	6	0	$\frac{1}{32}$	1/11.1	140
3	8	14	6	0	$\frac{3}{64}$	1/11.9	150
4	8	14	6	0	$\frac{1}{8}$	1/10.2	204
5	8	14	6	0	$\frac{3}{16}$	1/11.0	218
6	8	14	6	0	$\frac{1}{4}$	1/10.7	238
7	8	14	6	30	$\frac{3}{64}$	1/9.1	216
8	8	14	6	45	$\frac{3}{64}$	1/9.1	237
9	8	14	6	60	$\frac{3}{64}$	1/9.0	246
10	8	0	6	0	$\frac{3}{64}$	1/10.0	136
11	8	6	6	0	$\frac{3}{64}$	1/12.5	141
12	8	22	6	0	$\frac{3}{64}$	1/13.1	155
13	8	30	6	0	$\frac{3}{64}$	1/15.3	133
14	0	14	6	0	$\frac{3}{64}$	1/15.8	144
15	16	14	6	0	$\frac{3}{64}$	1/17.8	145

from the start of the cut up to the time of complete tool breakdown, and n is the slope of the curve or line as plotted on log-log paper. It represents the exponent of the tool life T. C is a constant which varies with the cutting conditions other than cutting speed and tool life. The length of tests was from 2 to 60 minutes, and it was found that all points within this range satisfied the tool-life equation and gave a straight line relation when plotted on log-log paper.

The Testing Equipment and the Set-Up

In these tests a heavy-duty lathe having a 30-inch swing and 14-foot bed of the geared head type was used. Power was supplied to the lathe driveshaft by a 15-horsepower, 40-degree, 60-cycle, alternating current motor driving through a No. 4 vertical type Reeves variable speed transmission. Silent chain drive was used from the motor to the Reeves and from the Reeves to the driveshaft of the lathe. By means of the change gears in the head of the lathe and the variable-speed transmission, the surface speed of the test log could be adjusted to any desired value. This speed was checked at all times with the surface speed indicator. The recorded cutting speed was measured while the tool was cutting. The depth of cut in all cases was checked with micrometers or calipers.

The tool bits were clamped into the tool holder by set screws as illustrated in Fig. 1. The tool holder, in turn, was held to the carriage by straps in the usual manner. The tool holder and bit

were set at right angles to the axis of the work, so that the side-cutting edge of the tool shown in Fig. 1 of 0 degrees makes a setting angle with the work of 90 degrees. Similarly, a 30-degree side angle tool had a setting angle of 60 degrees, etc. In other words, the setting angle was, in all cases, the complement of the side-cutting angle. In this way, the end-cutting angle of 6 degrees remained constant in all cases.

The tool life in minutes was measured by means of a stop watch, and was the time from the start of the cut until the tool failed. In some instances, the tool did not fail in one continuous cut, but the tool was started on the same depth and feed on the diameter left by the previous cut and run until failure would occur. Such interruptions did not show any interference or variation in the tests.

Tool failure was indicated by the sudden breaking down of the cutting edge. Tool failure has been found to be of two types, end and side failure. In end failure, the tip end of the tool seemed to burn out leaving a hollow which caused a marked decrease in the depth of the cut being taken. The more pointed tools fail by this method. Side failure occurs with tools of larger radii or smaller setting angles. These tools will still cut for some time after the first sign of failure has occurred without losing depth of cut or changing the character of the finish.

Tool wear takes place in progressive stages. For example, when the cutting speed is set for a 10-minute tool life for the 8-14-6-0-3/64-R tool, the wearing of the tool takes place somewhat as follows: at the start of the cut, the chip flows off the tool in a relatively straight and continuous ribbon. The built-up edge on the face of the tool is maximum in size. As the chip rubs over the built-up edge and the tool face, a groove or cup is gradually worn in the face. This causes the chips to curl and form helical or spiral coils, whose diameter is proportional to the amount of wear. The first signs of the grooving, as indicated by the form of chip, appear at the end of about one minute. At approximately five minutes, the diameters of the helical coils have become a minimum of about ¼ inch and the chips are broken up as they curl around and hit the test log. This condition continues until tool failure occurs. The cup formed in the tool wears larger, its edge approaching the tool cutting edge. At the same time, the flank of the tool is worn off slightly. The built-up edge under this condition becomes narrower.

When the worn surfaces of the groove and flank meet, a new irregular keen cutting edge is formed, after which tool failure soon occurs. The built-up edge is presumably non-existent under this condition and an optimum cutting tool shape is obtained except for tool endurance.

The color of the chips at different times indicates that the temperature is highest at the start of the cut with the newly ground tool, and that it decreases gradually to a minimum just before the irregular cutting edge is formed by the worn flank and cupped surface. The temperature again rises after this point, as the chip being removed is more irregular.

The above conditions have been confirmed by the use of the tool-work thermocouple in which the temperature between the tool and the work has been measured every one-quarter minute throughout the tool life. An interesting side light is that all of the three components of the cutting force follow, in general, the temperature curve. They are all maximum at the start of the cut and gradually reduce to a minimum just before the newly-formed cutting edge breaks down, after which they rise rapidly to high values. It has been found that high side-rake angles change this cutting condition. A report of the correlation of these pressures and temperatures throughout the tool life for this same set of tools will be given in a later paper.

Whenever a tool failed, the shoulder at the end of the cut was burnished. This burnished and work-hardened metal was always removed by a separate tool before starting another tool-life test. All cuts reported in this paper were made dry.

The Experimental Data

The experimental data are all based on a constant depth of cut of 0.100 inch and a constant feed of 0.0125 inch per revolution. All cuts were taken dry. The tool bits were ground in accordance with the variables listed in Table I. These data showing a relation between cutting speed and tool life for each set of variables, such as different nose radii, different side-cutting angles, different side-rake angles, and different back-rake angles, are presented and discussed separately below.

Effect of Variable Nose Radius—To find the influence of varying nose radius, a group of bits were ground in which the tool

angles were held constant as follows: back rake 8 degrees, side rake 14 degrees, end-cutting angle 6 degrees, side- and end-clearance angle 6 degrees, side-cutting 0 degrees, and the setting angle with the tool at right angles to the work 90 degrees. Four bits were first ground with a 0-inch radius, that is, a definite sharp point as left on the machine grinder. After hand grinding the radius on

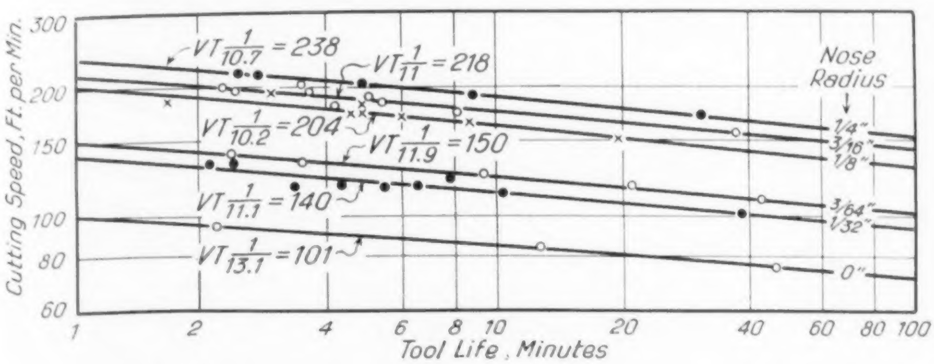

Fig. 2—The Effect of Nose Radius on the Performance of High Speed Steel Lathe Tools, 8-14-6-0-Variable R, When Cutting S.A.E. 2335 Steel Dry with 0.100 inch Depth of Cut and 0.0125 Inch Feed per Revolution.

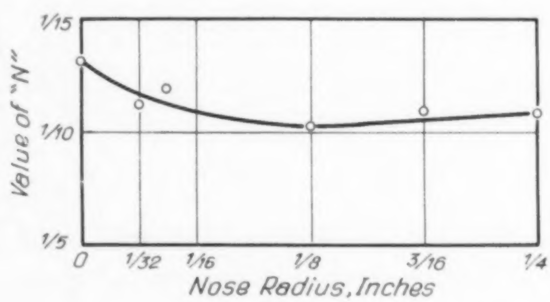

Fig. 3—Variation of (n) in the Equation $VT^n = C$, When Changing the Nose Radius of 8-14-6-0-Variable R Tool, Using High Speed Steel Tools Cutting Dry S.A.E. 2335 Steel at 0.100 Inch Depth of Cut and 0.0125 Inch Feed Per Revolution.

each bit to a templet, the flank was lightly honed. Sufficient duplicate tests were made to indicate that this slight honing had no appreciable influence on the life of the given tool. The tools were tested, reground with these same radii, and retested. With these tools, the life at various speeds was obtained as indicated by the lowest curve in Fig. 2.

The bits were subsequently reground with different radii as follows: $\frac{1}{32}$, $\frac{3}{64}$, $\frac{1}{8}$, $\frac{3}{16}$, and $\frac{1}{4}$ inch. Tool-life curves for each set of constant radii were obtained as shown plotted on log-log paper in

Fig. 2. The lines as drawn appear to be practically parallel. A close examination, however, shows that the slope, which corresponds to n, the exponent of T, actually changes from 1/13.1 for the 0-degree radius tool to 1/10.2 for the $\frac{1}{8}$-inch radius tool. The $\frac{1}{4}$-inch radius tool shows a value of n equal to 1/10.7, slightly larger than that for $\frac{1}{8}$-inch radius. These values are shown plotted in Fig. 3

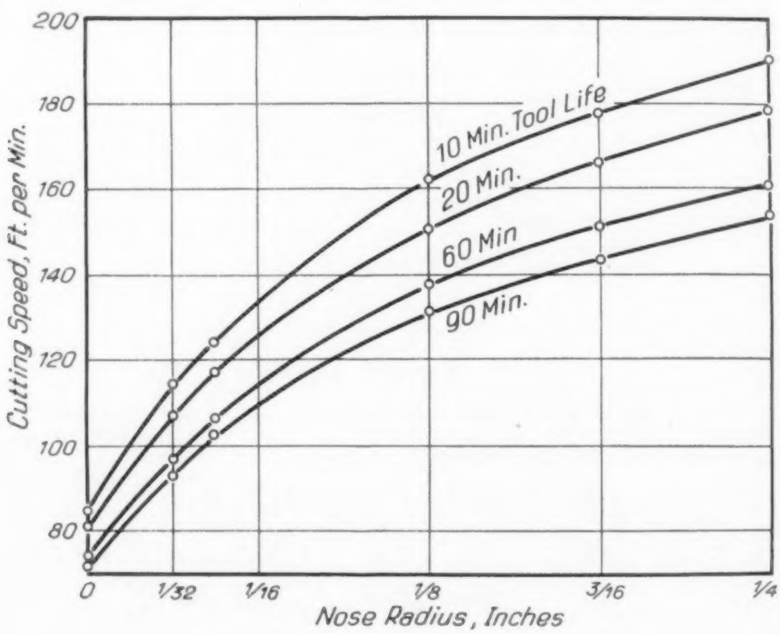

Fig. 4—The Effect of Nose Radius on Tool Life.

over the nose radius. The variation, except for the 0-degree tool, is apparently well within normal experimental error and seems to average about 1/11.

The influence of the nose radius on the permissible cutting speed for a definite tool life is illustrated in Fig. 4. The cutting speeds, as taken from Fig. 2, for a definite tool life of 10, 20, 60, and 90 minutes for each nose radius tool are shown graphically. It is seen that the radius of the nose has a very decided influence on the allowable cutting speed for a given tool life. From the 60-minute tool-life curve of Fig. 4 it is seen that a cutting speed for the 0-inch nose radius tool is 74 feet per minute. This increases to 97 for the $\frac{1}{32}$-inch radius tool, 106 for the $\frac{3}{64}$-inch tool, 138 for the $\frac{1}{8}$-inch tool, 151 for the $\frac{3}{16}$-inch tool, and 161 for the $\frac{1}{4}$-inch tool, showing a total overall increase from 74 to 161 feet per minute. Similarly, these curves show, for tools with any given nose

radius, the cutting speed for 10, 20, 60, and 90 minutes, respectively.

A mathematical relation between the nose radius in inches R, the cutting speed in feet per minute V, and the tool life in minutes T for the depth of cut of 0.100 inch and the feed 0.0125 inch, is obtained by first plotting the nose radius against the cutting speed

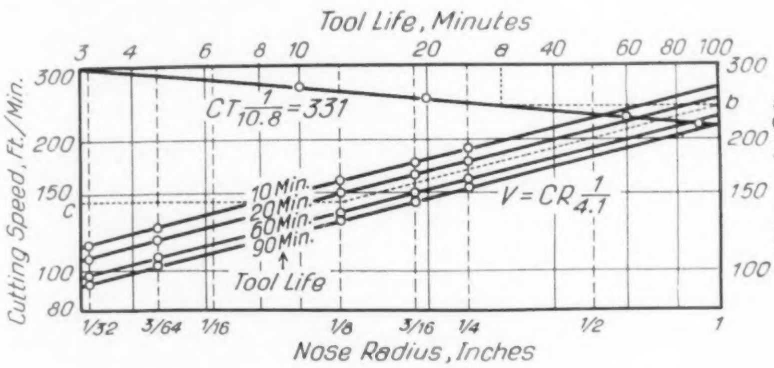

Fig. 5—Method of Obtaining Formulas Showing Relation Between Nose Radius, Tool Life, and Cutting Speed. High Speed Steel Cutting S.A.E. 2335, 0.100 Inch Depth by 0.0125 Inch Feed. 8-14-6-0-Variable R Tool. The Cutting Speed for a Tool Bit 3/8 Inch Square of High Speed Steel of the 8-14-6-0 Form Having any Value of Nose Radius, is Obtained for a 30-Minute Tool-Life from the Chart Above as Follows: The Point (a) Representing a 30-Minute Tool-Life on the Tool-Life Scale is Selected. The Dashed Line is Projected Vertically Until it Intersects the Curve for Tool-Life Independent of Nose Radius, Then Projected Horizontally to (b), Thence Parallel to the Four Parallel Lines Until it Meets the Vertical Line Passing Through the Desired Value of Nose Radius, Such as 1/8 Inch, Thence Horizontally to (c) on the Speed Scale.

in feet per minute on log-log paper. This gives four parallel lines as shown in Fig. 5, which resulted in the equation

$$V = C \times R^{1/4.1} \qquad \text{(IV)}$$

The value of *C* was found to be a direct function of the tool life by the equation

$$C \times T^{1/10.8} = K = 331 \qquad \text{(V)}$$

as shown in Fig. 5.

Combining equations (4) and (5), a final equation involving V, T, and R is obtained as follows:

$$V \times T^{1/10.8} = 331\ R^{1/4.1} \qquad \text{(VI)}$$

The graphical method for determining the cutting speed for a given tool-life of a tool of the 8-14-6-0 form is described in Fig. 5.

The finish left on the work by the tools of different radii va-

ried considerably. The 0-inch radius tool, as would be expected, left a very rough surface which could not be tolerated in most production work. When the tool had worn slightly, small chips or flakes of the test log were left on the surface and could not be removed, except with a file. As the nose radius was increased, the finish was improved until, with the $\frac{3}{16}$-inch radius tool, there was a slight tendency to chatter. This left its mark on the test log. The chatter was not heavy, however. With the $\frac{1}{4}$-inch radius tool, the chatter was so heavy as to be objectionable. A corrugated finish on the test log resulted. The chips produced had rough edges much like a circular saw tooth. The spacing of these teeth was uniform. Undoubtedly, this objectionable chatter accounts for the lower value of the exponent in the cutting-speed tool-life formula referred to above.

In summarizing the influence of varying radius of the tool on its performance, it is seen that the cutting speed for a given tool life may be increased by increasing the radius. For a 20-minute tool life, the increase of speed, as the nose radius is changed from 0 to $\frac{1}{4}$ inch, is 122 per cent. For a 60-minute tool life, the increase is about 118 per cent.

The most desirable cutting condition would be obtained with a radius of approximately $\frac{1}{8}$ inch. This tool gave a good finish, high cutting speed, and showed no tendency to chatter.

When the nose radius increases, the average thickness of chip decreases. Thin chips have a tendency to cause chatter and, with the $\frac{1}{4}$-inch radius tool where the chip was thinnest, the chatter was greatest. A wide curved chip, as formed by a radius tool, has a thickness that varies at different points along the cutting edge, and therefore, tends to cause numerous periods of vibration, any one of which would tend to neutralize another. Undoubtedly, this influence is a function of the ratio of the nose radius to the depth of cut. In this work, the depth of cut was 0.100 inch. The $\frac{1}{8}$-inch nose radius tool was found to cut smoothly with no tendency to chatter. When the radius was increased so as to become much larger than the depth of cut, as in the case of the $\frac{3}{16}$ and $\frac{1}{4}$-inch radius tools, the variation in thickness of chip was not sufficient to dampen the vibration.

The increase in the radius allows a longer cutting edge of the tool to do the work, thus distributing the pressure over a greater portion of the tool and providing a greater amount of tool material to

conduct away the heat generated. This results in higher cutting speeds for a given tool life. The increase in cutting speed, however,

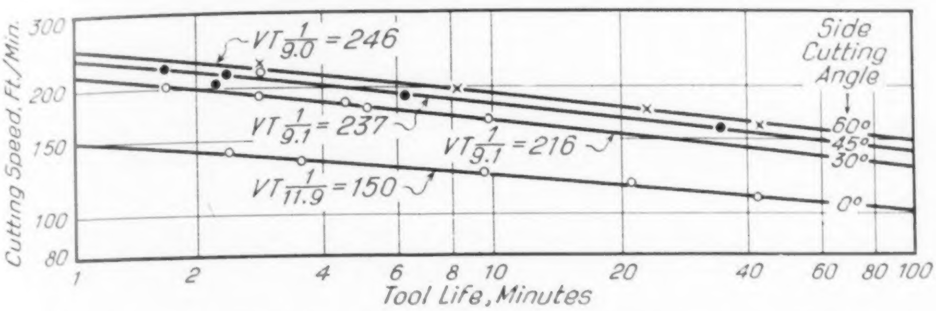

Fig. 6—Effect of Side-Cutting Angle on Tool Life and Cutting Speed When Cutting Dry S.A.E. 2335 Steel Using 8-14-6-Variable Side-Cutting Angle—$\frac{3}{64}$-Inch R High Speed Steel Lathe Tool with 0.100 Inch Depth of Cut and 0.0125 Inch Feed Per Revolution.

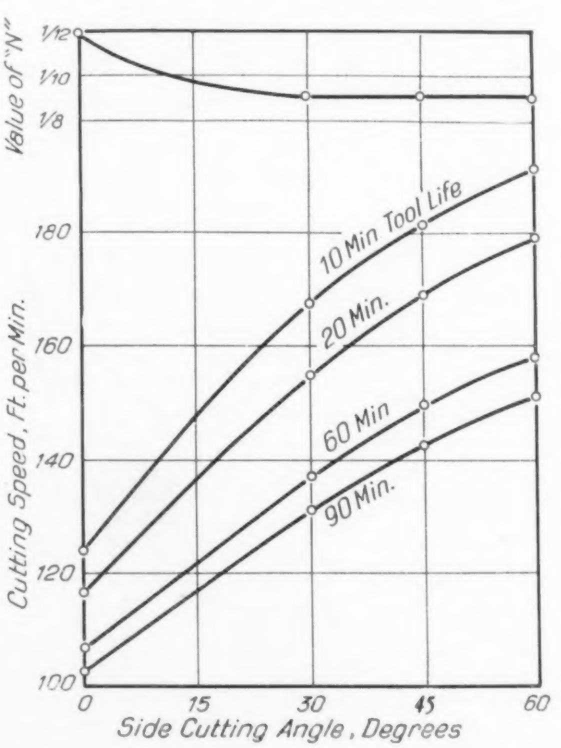

Fig. 7—The Effect of Varying Side-Cutting Angle on the Performance of a High Speed Steel Lathe Tool, 8-14-6-Variable Side-Cutting Angle-$\frac{3}{64}$-Inch R, When Cutting Dry an S.A.E. 2335 Steel with 0.100 Inch Depth of Cut and 0.0125 Inch Feed Per Revolution.

is not proportional to the increase in length of cutting edge. This is borne out in Fig. 4.

Effect of Variable Side-Cutting Angle—In studying the influ-

ence of the side-cutting angle, tools were ground with side-cutting angles of 0, 30, 45, and 60 degrees, respectively, with all other angles constant, such as back rake 8 degrees, side rake 14 degrees, end and side clearance 6 degrees, end cutting 6 degrees, and nose radius $\frac{3}{64}$ inch. The experimental data for these four sets of tools are shown plotted on log-log paper in Fig. 6. Four straight lines are shown with their respective equations and constants. The exponents n for the 30-, 45-, and 60-degree tools are practically identical at 1/9, while the exponent n for the 0-degree side-cutting angle tool is 1/11.9. It is seen further that the constants for those tools having the greatest side-cutting angle are greatest.

A relation between the side-cutting angle and the cutting speed for several specific values of tool life is shown in Fig. 7. For a 60-minute tool-life, the cutting speeds for 0-, 30-, 45-, and 60-degree side-cutting angle tools are 107, 137, 149, and 158 respectively. Expressed in relative values, these speeds become 1.00, 1.28, 1.39, and 1.48, respectively.

Allowable cutting speeds for 10-, 20-, 60-, and 90-minute tool-life also are shown for the various side-cutting angles in Fig. 7.

The effect of changing the side-cutting angle on the cutting speed when the depth of cut and feed are constant may be represented by the following equation

$$\frac{V}{(a + 15°)^{1/3.8}} = C \qquad \text{(VII)}$$

in which a is the side-cutting angle in degrees and C, a constant, is a function of the tool life. This equation is developed from the four parallel lines expressing a relation between the cutting speed as ordinates and side-cutting angles as abscissas, as shown in Fig. 8.

By plotting the constant C, determined from each of the four straight lines in Fig. 8, as ordinates over the cutting times of 10, 20, 60, and 90 minutes respectively, as abscissas, C is found as a function of tool life T, as follows

$$C \times T^{1/9.1} = K = 78 \qquad \text{(VIII)}$$

Substituting the value of C in this equation in that of equation (VII), the following is obtained

$$\frac{V\,T^{1/9.1}}{(a + 15°)^{1/3.8}} = 78 \qquad \text{(IX)}$$

It is seen that the two points farthest to the left on the lowest two lines of Fig. 8 are slightly out of place. These points represent the 0-degree side-cutting angle plus the constant of 15 degrees. This error is due to the fact that the exponent of the tool life in the cutting-speed tool-life equation is 1/10.7, rather than 1/9.1, which obtained for the balance of the tools.

The quality of finish was not noticeably affected by a change in the side-cutting angle, except where chatter occurred with the side-cutting angles of 45 degrees or more, in which case a corru-

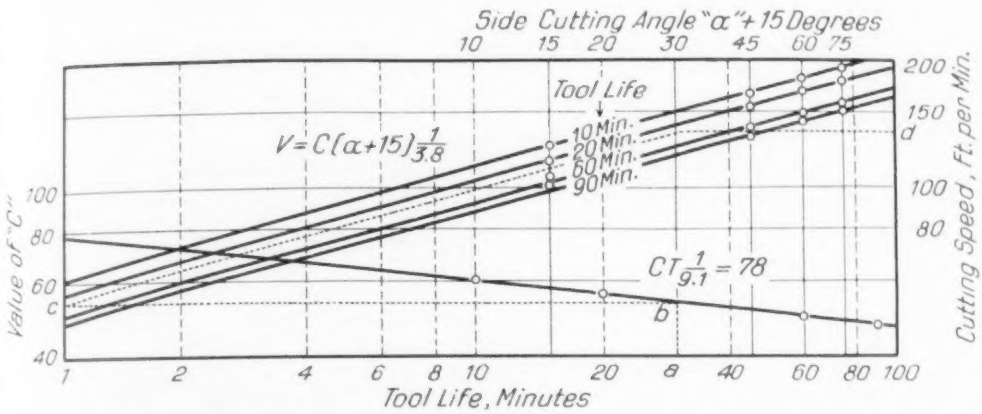

Fig. 8—Method of Obtaining Formula Showing Relation Between the Side-Cutting Angle (α), Tool-Life, and Cutting Speed, When Using 8-14-6-Variable Side-Cutting Angle-$\frac{3}{64}$ Inch R High Speed Steel Lathe Tool Cutting S.A.E. 2335 Steel Dry with 0.100 Inch Depth of Cut and 0.0125 Inch Feed Per Revolution. Example: To Find the Cutting Speed Under the Stated Conditions for a 30-Minute Tool-Life of a 15-Degree Side-Cutting Angle Tool, Follow the Dashed Line from (a) to (b), (c) and (d).

gated surface resulted. For the 45-degree tool, chattering was great enough to be objectionable. For the 60-degree tool, chatter was intense and formed saw-edged chips.

In summarizing the influence of the side-cutting angle, the following points were noted. By increasing the value of side-cutting angle, an increase in cutting speed for a 20-minute tool-life is obtained for the 0-, 30-, 45-, and 60-degree side-cutting angles in proportion to 1, 1.33, 1.45, and 1.54. This corresponds to the relative values of tool life for a cutting speed of 117 feet per minute, which gives a tool life of 20 minutes to the 0-degree side-cutting angle tool, of 1, 13, 28, and 48. The value of the exponent n is 1/9.1 for the tools with side-cutting angles of 30, 45, and 60 degrees, but is decreased to 1/11.9 for the 0-degree side-cutting angle tool, as shown in Fig. 7.

Effect of Variable Side-Rake Angle—In determining the in-

fluence of the side rake angle, a series of tools having a back rake angle of 8 degrees, an end-cutting angle of 6 degrees, side- and end-clearance angles of 6 degrees, side-cutting angle of 0 degrees, and a nose radius of $\frac{3}{64}$ inch, was ground successively with 0-, 6-, 14-, 22-, and 30-degree side rake angles. Tests were run with each group separately. The depth of cut was 0.100 inch, the feed was 0.0125 inch, and the cutting was done dry.

The data resulting from these tests are shown plotted on log-log paper in Fig. 9. Five straight lines are obtained, one for each set of tools of a given side-rake angle. The cutting-speed tool-life

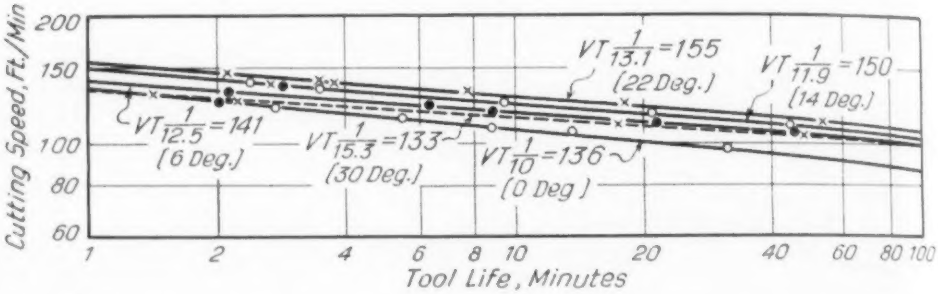

Fig. 9—Effect of Varying Side Rake Angle on Tool-Life and Cutting Speed When Using 8-Variable Side-Rake -6-0-$\frac{3}{64}$-Inch R High Speed Steel Lathe Tool, Cutting S.A.E. 2335 Steel with 0.100 Inch Depth of Cut and 0.0125 Inch Feed Per Revolution.

equation for each series of tools is indicated and the values of n and C are given. It is seen that the value of n is not constant for all groups of tools. The value of n for the 0-degree side rake angle is $\frac{1}{10}$. This value decreases to 1/15.3 for the 30 degree side rake angle, as shown also by the upper curve in Fig. 10. It is fairly uniform for intermediate values of side rake angle.

The effect of this change of n is important in deciding the best side-rake angle to be used. When a short tool life of approximately 20 minutes is to be used as a basis of tool comparison, the side rake angle which will give the highest cutting speed is 20 degrees, as shown in Fig. 10. If a tool-life up to 300 minutes is desired, those tools with a side rake of from 23 to 24 degrees are shown to be superior. For general use, however, for the cutting conditions as specified in these tests, the 22-degree side rake angle appears most favorable.

For a 20-minute tool-life, the allowable cutting speed for the various side rake angles of 0, 6, 14, 22, and 30 degrees will be 101, 111.5, 118, 122, and 110, respectively. These cutting speeds may be

represented by ratios of 1, 1.10, 1.17, 1.21, and 1.09, respectively. The values of tool life for the cutting speed which will give a 20-minute tool-life for the 0-degrees side rake tool may be expressed as 1, 3.8, 6.1, 12.0, and 3.8. The finish left on the test log was somewhat better for the larger values of side rake, although the variation was small. Chatter was not present in any of the tests.

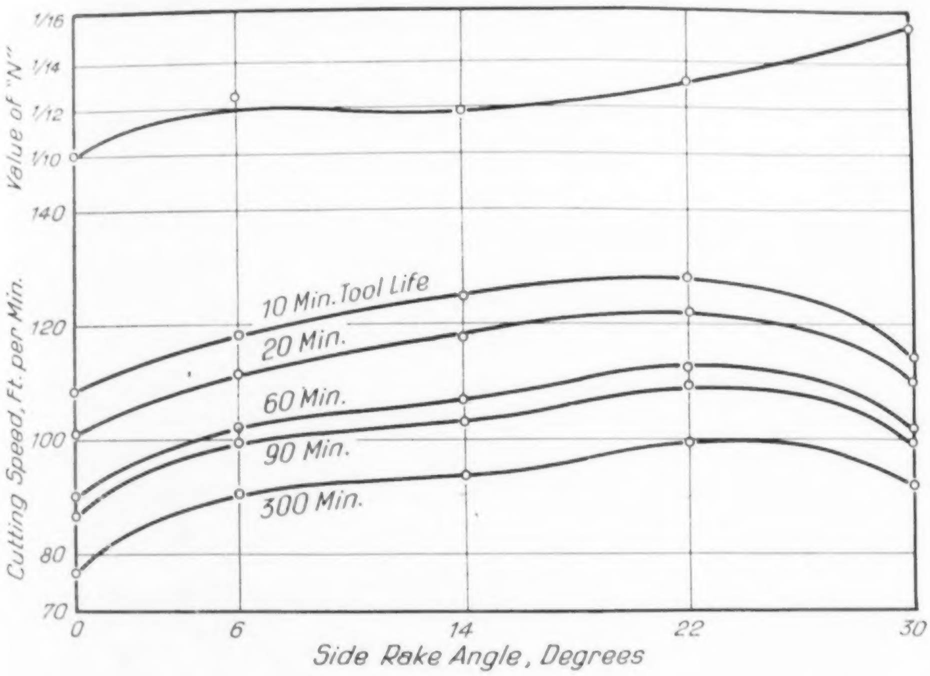

Fig. 10—Effect of Side Rake Angle on Performance of High Speed Steel Lathe Tools, 8-Variable Side-Rake-6-0-$\frac{3}{64}$ Inch R, When Cutting Dry an S.A.E. 2335 Steel with 0.100 Inch Depth of Cut and 0.0125 Inch Feed Per Revolution.

For side rake angles of 0 and 6 degrees, the chips were well curled as they came from the tool and broke into short lengths which were easily disposed of. When side rake angles of 22 and 30 degrees were used, the chips were long and straight, and curled very little until just before tool failure. These chips were bothersome and had to be pulled away from the machine in order to make it possible to observe the cutting action of the tool.

For the 30-degree side rake angle tools, there was practically no cup worn in the tool face by the rubbing action of the chip, and the tool failed by the rounding off of the cutting edge.

Effect of Variable Back-Rake Angle—Three separate sets of tools were used in studying the influence of back rake angles. Back

rake angles of 0, 8, and 16 degrees were used successively. The constant angles of the tool were side rake 14 degrees, end cutting 6 degrees, front and side clearance 6 degrees, side cutting 0 degrees,

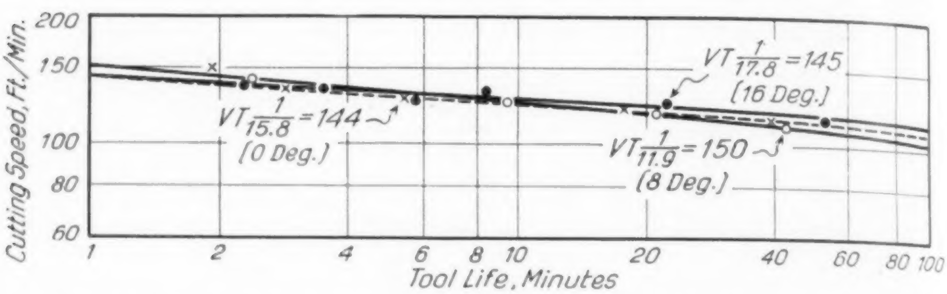

Fig. 11—Effect of Back Rake on Performance of High Speed Steel Lathe Tools, Variable Back Rake-14-6-0-$\frac{3}{64}$ Inch R, when Cutting an S.A.E. 2335 Steel Dry with 0.100 Inch Depth of Cut and 0.0125 Inch Feed Per Revolution.

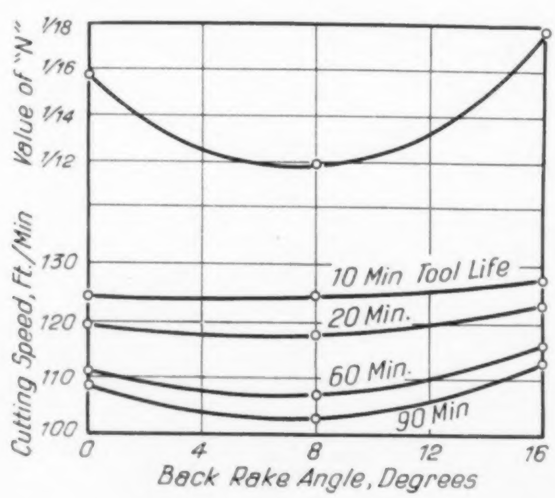

Fig. 12—Effect of Back Rake Angle on the Performance of High Speed Steel Tools, Variable Back Rake-14-6-0-$\frac{3}{64}$ Inch R, When Cutting Dry an S.A.E. 2335 Steel With 0.100 Inch Depth of Cut and 0.0125 Inch Feed Per Revolution.

and nose radius $\frac{3}{64}$ inch. The results of the experiments are shown plotted on log-log paper in Fig. 11.

It appears that the most important influence of the back rake angle is to change the value of n. This rotates the curves about a point between the 4- and 12-minute tool-life. The slope of the 0-degree back rake tool is less than that of the 8-degree back rake tool, but greater than that of the 16-degree tool. Values of n as shown in Fig. 11 are replotted in Fig. 12. From Fig. 11 it appears that the highest cutting speed for any tool life up to 6 minutes can

be obtained with the 8-degree back rake tool. For a tool life of more than 10 minutes, the tool with the 16-degree back rake gives the greatest cutting speed for a given tool-life, with the 0-degree and 8-degree back rake angle tools next in order, as shown in Fig. 12. The authors have found in other similar tests using a 30-degree side-cutting angle in place of the 0-degree, that the 8-degree back rake angle tool shows up to best advantage for all values of tool life up to 90 minutes. The finish obtained was better for the higher values of back rake angle, but the variation was relatively small.

Check Tests—The results presented in this report have been checked by making two individual series of tests with the same tool material when covering the same range of tool angles. In the tests reported, a basic tool with a 0-degree side-cutting angle was used. In the first check test on the same material, a 30-degree side-cutting angle tool was used. A second check test, using the same tool material and the same angles as used for the tools in this report, but with a test log of 0.61 per cent carbon steel, also was run. In all cases, however, the change in tool angles produced the same general change of the values of n and tool-life. The results were consistent and could be reproduced at any time.

Conclusions

Running all cutting tests dry with a depth of cut of 0.100 inch and a feed of 0.0125 inch when cutting an S.A.E. 2335 steel with ⅜-inch square high speed steel tool bits of the regular 18-4-1 type, the following results were observed when variables of nose radius, side-cutting angle, side rake angle, and back rake angle were introduced.

1. When the nose radius was varied from 0 to ¼ inch, the value of n in the cutting-speed tool-life equation $VT^n = C$ was found to be 1/13.1 for the sharp-pointed tool, but about 1/11 for all other tools, as shown in Figs. 2 and 3.

2. For this size of cut, the $\frac{3}{16}$-inch nose radius tool chattered occasionally, but the ¼-inch radius tool chattered badly. The ⅛-inch nose radius tool cut best considering the high value of the constant and the vibrationless cutting.

3. A definite relation is found to exist between nose radius, tool-life, and cutting speed, as shown in Fig. 5.

4. With the tool-life range up to 100 minutes, the constant

C of the cutting-speed tool-life equation is approximately proportional to the value of the nose radius, that is, the greatest constant is obtained with the greatest value of nose radius. See Fig. 2.

5. The value of n for a 0-degree side-cutting angle is 1/11.9, but for the 30-, 45-, and 60-degree angles it is 1/9. See Figs. 6 and 7.

6. The constant in the cutting-speed tool-life equation increases as the side-cutting angle is increased. A much greater tool-life is obtained for the large side-cutting angles. Chatter is apt to occur, however, for side-cutting angles greater than 45 degrees. See Figs. 6 and 7.

7. A definite relation exists between the side-cutting angle, tool-life, and cutting speed, as shown in Fig. 8.

8. The side rake angle influences the value of n. The value of n decreases as the side rake increases. See Figs. 9 and 10.

9. The constant in the cutting-speed tool-life equation increases up to a limiting value of side rake angle, after which the constant falls off. See Fig. 9.

10. The value of n changes with a change in back rake. The maximum value of n is obtained in these tests at approximately 8 degrees back rake. See Figs. 11 and 12.

11. For long tool-life, the 16-degree back rake angle gives the greatest allowable cutting speed, but for short tool-life, the 8-degree back rake angle is best. The 0-degree back rake angle is better than the 8-degree angle for long tool life. See Fig. 11.

12. It has been found that the size of the cut influences the value of n. The lighter feeds give the lowest values. This is an indirect conclusion from the tests run, but is believed to explain why the value of n, as determined in this report, varies from 1/9 to 1/17.8, while that of French and Digges was 1/7 and that of Taylor was $\frac{1}{8}$ for heavier cuts with high speed steel tools.

Acknowledgments

The authors wish to express their appreciation to Messrs. C. E. Kraus, W. L. Weiss, and Vasily Prianishnikoff for their assistance in carrying out the tests. The authors are indebted to A. F. MacFarland, of the Cyclops Steel Co., Detroit, Michigan, who furnished gratis the tool bits for the tests; to M. F. Judkins of the Firth-Sterling Steel Co., McKeesport, Pennsylvania, who furnished gratis the

2000-pound 0.61 per cent carbon test log; and to Coleman Sellers, 3rd., of William Sellers Company, Philadelphia, for machine grinding all tool bits for the first tests.

The Graduate School of the University of Michigan very generously provided Mr. Gilbert with a research fellowship so that this work might be carried out.

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DISCUSSION

Written Discussion: By T. G. Digges, associate metallurgist, Bureau of Standards, Washington, D. C.

This interesting paper by Boston and Gilbert is a welcome contribution toward a clearer understanding of the factors that influence the cutting speed or tool life of high speed steel lathe tools.

With reference to the data on the series of tests in which the nose radius was the variable, the authors find that the slope of the lines representing the relation between cutting speed and tool life varied, in the logarithmic plotting used in Fig. 2, but conclude that "the variation, except for the 0-degree tool, is apparently within normal experimental error and seems to average about $\frac{1"}{11}$. The writer would like to call attention to this statement because of the large number of variables encountered in tool testing, particularly with shallow cuts. Appreciable variations are encountered in the cutting properties from point to point in large test forging and only minor variations from the intended depth of cut represent a large percentage variation with shallow cuts and have a relatively marked effect upon the observed tool life so that reproduction of results becomes difficult.

The variation in the nose radius of the tool showed a marked effect on the value of C in the equation $VT^n = C$, ranging from about 100 with 0 inch nose radius to about 240 for ¼ inch nose radius. Thus, the change in tool life with

nose radius, as given in Fig. 4, is due to this variation in "C" in the above equation and not to the exponent "n."

The writer believes that it should be pointed out that the relations involving V, T and R as given in equation (VI) are applicable only to test conditions similar to those used in the present experiments until further experimental evidence is obtained. This equation would probably require considerable modification to be applicable, for example, to the conditions of roughing cuts or to larger size tools.

Tests made at the Bureau of Standards and reported in Research Paper No. 120, Bureau of Standards Journal of Research, Vol. 3, 1929, p. 829, with ¼-inch by ½-inch high speed steel tools with different back rake and side rake angles, which were tested under similar conditions as regards cutting material, feed and depth of cut as those used by the authors, also showed that the tool angles affected the performance of tools to an appreciable degree with shallow cuts. The cutting speed for a 90-minute tool life increased with back rake to 30 degrees and decreased with further increase in back rake. Also the tool life was greater in the tests made with tools of 0-degree side rake than with the tests made with the tools having 8-degree side rake.

On page 562, the authors have made reference to a tool life of 300 minutes. It is indeed questionable if the equations used to represent the experimental data are applicable for such long time cutting. The writer has found it difficult to determine the failure of high speed steel tools with shallow cuts in tests of this duration without some method of determining the end point other than the glaze on the forging.

In the lathe tests made at the Bureau of Standards with high speed steel tools under shallow cuts and without cutting liquids, the value of n varied from 1/10 to 1/12. These values are in close agreement with those of the present authors when the cutting conditions are similar.

The authors state that "it has been found that the size of the cut influences the value of n. The lighter feeds give the lowest values. This is an indirect conclusion from the tests run, but is believed to explain why the value of n, as determined in this report, varied from 1/9 to 1/17.8, while that of French and Digges was 1/7 and that of Taylor was 1/8 for heavier cuts with high speed steel tools."

Experimental data do indicate that the size of cut influences the value of n. The results given in the present report, therefore, should be compared with the tests made under similar cutting conditions, not with heavy cuts. The writer does not believe that the variation of n from 1/9 to 1/17.8, as given in this report, can properly be attributed to a change in the size of the cuts, when it is stated that all tests were made at a constant cut of 0.100 inch depth and 0.0125 inch feed per revolution. It appears more probable that the variation is to be correlated with the change in tool shape.

Written Discussion: By Alvan L. Davis, research engineer, Scovill Manufacturing Company, Waterbury, Conn.

Our knowledge of machining metals is steadily expanding, through the unflagging efforts of Professor Boston, his co-workers, and other investigators. The present paper gives us one cross section; the one which applies to turn-

ing forged steel S.A.E. 2335 (C = 0.36 per cent) in what may be called a semi-annealed condition.

Several of the interesting conclusions of the authors are checked as correct by a very skilled lathe hand, with whom I discussed the results. This man's choice of shape for a ⅜-inch tool bit for turning S.A.E. 1035 is as follows:

Nose radius	$\frac{1}{16}$ inch
Side angle of tool	15 degrees
Side rake of tool	22 degrees
Back rake of tool	little or none
Clearance angles at side and end	6 or 7 degrees

In general, he says, "have nose radius and side angle as large as can be without causing chatter, and without getting excessive thrust on tool."

For cutting soft machinery steel, or brass, he would increase clearance angles from 6 or 7 up to 12 to 18 degrees, also side rake angle up to 30 degrees.

Written Discussion: By Lennox F. Armstrong, president, Armstrong Bros. Tool Co., Chicago.

There is no doubt that Professor Boston's work is outstanding particularly from the point of correlating former data which we have available to the present marked trend toward alloy steels. This paper and the graphs presented probably mark the initiation of further extensive work in the field of correct cutting angles and basic formulæ for the determination of same. The angles and cutting procedure suggestions in each case fix in a definite table form an experience over a number of years, and I personally feel very much pleased that Professor Boston's data correspond so closely to what we as tool manufacturers have felt were average correct cutting angles. Of course the discovery that the formula $V\ T^n = C$ applies to the cutting of high strength alloy steels with our modern 18-4-1 high speed steels is a stride forward in a field in which present scientific knowledge is very meager.

Written Discussion: By G. C. Riegel, chief metallurgist, Caterpillar Tractor Company, Peoria, Ill.

We have not done sufficient work, either confirmatory or otherwise, to speak with any authority on the splendid treatment of this subject as accorded by Professor Boston. There are, however, a few questions which occurred to us in reading over this article which we would like to have referred to Professor Boston in advance of his presentation of his paper.

It is observed, in both the abstract and also in the last paragraph, page 549, the statement that S.A.E. 2345 steel was used for the test, whereas on the top of page 550, the chemistry for this analysis was given as C 0.36, P 0.32, S 0.36. We assume that these are typographical errors in view of the fact that 0.36 carbon would not be classed as S.A.E. 2335, and both the sulphur and phosphorus figures are undoubtedly improperly written.

In the paragraph at the bottom of page 552 and top of page 553, while it is stated that the tool bit and holder were set at right angles to the axis of the work, we failed to observe a statement with respect to whether the tool bit was always in a horizontal plane with the axis of the work, above or below. If there were any deflection of either the tool holder or the tool point under the conditions of feed or speed, no mention was made of this fact, that is the rigidity of the tool holder.

While not a part of this experiment, it no doubt was recorded by Professor Boston, and would be information of value, that is, the power consumption required to cut the several types of steel used in these experiments.

In conclusion, it must be said that Professor Boston is placing all of us engaged in machining metals under grateful obligation for the splendid researches which he has promoted and published on this and kindred subjects.

Written Discussion: By Coleman Sellers, 3rd, William Sellers and Company, Inc., Philadelphia.

The authors are to be congratulated on their very carefully carried out tests and comprehensive presentation of data and discussion of results. This is certainly one of the most valuable papers presented in the past decade on the subject of cutting of metals. It is to be hoped that the investigation will be carried further to include different depths of cut, different feeds and different materials of work and tool. In general the trends will probably be as indicated except in the case of the nose radius but will the exponent remain the same? The constant, of course, will change largely on account of the greater heating with increases in amount of metal removed during the life of the tool.

As the authors have pointed out, there has long been a great deal of confusion as to the relation between cutting speed and tool life per grind. There has also been much confusion as to the correct shape of tool to use in roughing. Taylor's experiments led him to adopt and recommend his standard round nose tool with certain clearances and angles. Many plants adopted his shape but its use was by no means universal due, probably, to the fact that it was difficult to reproduce without the proper type of tool grinding machine. The latter did not readily fit into the method used in most shops of permitting every man to grind his own tools. To attempt to have machine operators grind by hand to the correct shape seemed out of the question and so other shapes were used.

It would be exceedingly interesting to find out how Taylor's shape of tool compares with other shapes in common use today. It is quite likely that a more efficient general purpose tool can be or has been developed for modern cutting materials. In the twenty-eight years or so since his famous paper there has been great improvement in high speed steels, and other materials such as Stellite and tungsten carbide have been introduced.

There is no doubt that better tools can be developed for a given set of conditions but they are applicable only in production work.

There have been very few data published on the subject of tool shape. Experimental work such as the authors have conducted is very helpful in arriving at a solution of this problem.

The angles of side and back rake determined as best by the authors agree with Taylor's angles. The side cutting angle and nose radius do not agree. One reason for the difference in nose radius is that Taylor did not consider any tool as small as ⅜-inch square. Furthermore, as his tool was to meet varying conditions he varied the side cutting angle which resulted in a curve. This curve of course had the added advantage as the authors have pointed out of giving numerous periods of vibration tending to neutralize each other.

The following table gives the actual size of Taylor's round nose tools for

roughing soft and hard material. The tools were intended for general purpose work. On the drawing are a series of lines representing depths of cut of $\frac{1}{16}$-inch, $\frac{1}{8}$-inch, $\frac{3}{16}$-inch, etc. Where these intersect the cutting edge the angle is given. These are shown in the table. This is the setting angle. It will be noted that it varies from approximately 41 degrees with $\frac{1}{16}$-inch depth to 79 degrees with $\frac{1}{2}$-inch depth for hard materials and approximately from 48 degrees with $\frac{1}{16}$-inch depth to 77 degrees with $\frac{1}{2}$-inch depth. These angles correspond to the following side cutting angles disregarding the fractions.

	Side Cutting Angle		Nose Curvature	
Depth of Cut	Hard Cutting Materials	Soft Cutting Materials	Soft Cutting Materials	Hard Cutting Materials
$\frac{1}{16}$	49 deg.	42 deg.	47° 45′	41° 5′
$\frac{1}{8}$	30	34	56° 20′	60° 5′
$\frac{3}{16}$	20	30	60° 0′	69° 58′
$\frac{1}{4}$	19	26	64° 0′	71° 35′
$\frac{3}{8}$	15	20	70° 35′	75° 15′
$\frac{1}{2}$	11	13	77° 15′	78° 57′

These will be seen to be quite different for the two tools.

Special tools of odd rakes and clearances are often required for certain operations. For instance, in planing soft structural steel, to obtain a smooth surface suitable for a flat bearing, the following shape was used for a 2 x 1-inch tool:

Top side rake .... minus 52 degrees
Setting angle .... 90 degrees
Clearance .... side, 3 degrees
Clearance .... front, 10 degrees
Lip radius .... 13/64 degree

This was a shear cut tool with a negative top side rake. That is, the inclination is opposite to that usually used.

Another special operation is turning the treads of used car wheels from passenger cars. These wheels are very hard being made of high manganese steel and cold-worked from pounding over the tracks for thousands of miles. The tool used for roughing is 3x1 inch section with no side or back top rake and with $\frac{5}{8}$ inch radius of nose. The depth of cut is usually from $\frac{1}{4}$ to $\frac{3}{8}$ inch and the feed is $\frac{3}{8}$ inch per revolution.

Written Discussion: By A. Wallichs, Professor, Technische Hochschule, Aachen, Germany.

The authors have used as the basis of their investigation the criterion of tool life, which has already been used as a basis for all machinability investigations since the first experiments by Taylor, and again and again it has proved to be successful by the majority of the later investigators. The expansion of these experiments towards a greater variety of cutting speeds leads to the development of tool life curves, through whose character and position one can easily come to interesting conclusions and practice. All developed tool life curves show straight lines in the double logarithmic system, which ran almost entirely parallel for the different cutting conditions. The authors have shown the curves not parallel in a few cases, but if further tests were to be made, then the parallel relation should prove correct. Earlier investigations

proved the correctness of the parallelism of the *TV* straight line on the double logarithmic system.

The authors show the importance of the nose radius which already has been shown by Taylor to be especially important. It is found that with a tool of too large radius the machine gets into vibration and so-called chatter occurs. This was not found by Taylor on roughing cuts. Some time ago Dempster Smith and I made experiments to find the influence of the setting angle. The factors resulting from the different setting angles as reported differ somewhat from our experiments.

It should be noticed that the authors investigated only one material whereas by the investigations of D. Smith and myself the average was taken for different kinds of materials, such as groups of steel and steel castings, on one hand, and cast iron groups on the other hand.

The other results relating to the importance of the rake angles correspond, as might be expected, with the measurements made in our institute.

Furthermore, the investigators deal in the determination of the coefficient n in the Taylor formula and their results correspond with those of earlier investigators.

It is noticed that there is a marked deviation in values of n for light cuts which also depends on different factors. I report that there were some variations of the exponent n in the formula ($VT^n = C$) in my investigations. For the heavy cuts the values vary according to the material between 1/8 and 1/13; a great number of the alloyed steels (Baustahl) show the value n equal to 1/9. Lower values, even down to 1/17, were found for light cuts as used on automatic lathes.

In consideration of the existing variations it is recommended that the straight-line double logarithmic system T-v be used for the proper representation in graphic form, and to mark the test points plainly on the diagram, because the formula easily leads to the belief of a fixed validity of the law.

William H. Oldacre:[1] As one interested in the production and application of cutting fluids I want to express our great appreciation to Professor Boston and his associates for the work that has been done in the investigation of the machining of metals. Frequently, in our contacts with industry, we have been impressed by the fact that plants having elaborate equipment and personnel for the control of the structure and the machinability of metals have wholly turned over to relatively uneducated and unskilled individuals the final process in that machining. It is not at all uncommon for plants to depend upon the observation of the machine operator for their final conclusions as to whether their preparation of the metal and their control of its machinability have been successful.

We feel frankly that if Professor Boston has done nothing more than point out the fact that this problem of machining is worthy of the attention of the metallurgical intelligentsia he has done us a great favor, and one which we greatly appreciate.

Haakon Styri:[2] I would like to ask Professor Boston if he would tell

[1]Director of research and engineering, D. A. Stuart and Company, Chicago.

[2]Director of research, S. K. F. Industries, Philadelphia.

what would be expected if we increase the nose radius still farther than in his paper.

Charles Kraus:[3] I have been connected with Professor Boston and Mr. Gilbert in this work for some time, and there are a few questions I would like to ask, not about the paper, but rather of the audience concerning some of the work that we are doing now.

The size of cut that was used in this paper, as stated in the paper, was one-tenth inch deep and 0.0125 inch feed.

We are doing considerable further work, and if we increase the size of the cut it uses considerable material, so naturally we are interested in keeping our size of cut relatively small. I would like to get some discussion, if possible, as to the most desirable size of cut for this work.

If we decrease the size of the cut too much it might not be of so much practical advantage, and if we increase it too much it uses too much material.

Also, in connection with this work we have been running some tests to determine what happens during the life of the tool as far as forces and temperatures are concerned. We have numerous graphs prepared to show the influence of these variables as a function of the time.

At present, our temperature is recorded in millivolts on a tool-worked thermocouple and in attempting to calibrate this thermocouple we find that we will get one rather smooth curve going up in the temperature scale, but the curve shows a hysteresis sort of effect in coming down. In other words, the curve is displaced vertically. The question I would like to bring up is how we can get this curve more accurate, whether by going up a little way and then coming down, and then going up farther and coming down. Has anyone had experience in the determination of these curves?

We ran up to a temperature of 1400 degrees Fahr. and consequently there were some changes in the properties of both the high speed steel and the sample of the material that we were using. We have at present both pieces of high speed steel rehardened and will be cutting new specimens of the material in the course of the tool life tests. These temperatures run above 1100 to 1200 degrees Fahr. at the point of cutting, and at failure they run up above 1400 degrees Fahr. for this size of cut.

If any of you have in mind any information that we could use in this work, we would very much appreciate your letting us know about it.

C. B. Sadtler:[4] I might possibly be able to offer a suggestion that would be of some help to Professor Boston and his associates, particularly in answer to the first part of the discussion of the previous speaker. He pointed out the difficulty of keeping the cost of testing within bounds, which is the thing that in these days concerns us more than it has in times past.

In respect to the large amount of test material that was consumed in the cutting, a great many have been confronted with that same thing. We have found that we not only could obtain economy in the quantity of material used but could answer in a practical way the question that is becoming more and more of concern to cutter manufacturers; that is, how to make cutters cut some of the harder structural materials.

[3]Instructor, University of Michigan, Ann Arbor, Mich.

[4]Metallurgist, Barber-Colman Company, Rockford, Ill.

Automobile, airplane and similar applications have necessitated the use of higher tensile material for spline shafts and gears, etc., and it is necessary to machine large quantities of such material economically. So that from the point of view of getting test information, if you will select the material heat treated running the Brinell up to say between 300 and 400, as compared with 207 which I noted is reported in Professor Boston's paper, you can cut down the amount of material you use very considerably and not necessarily make it impossible to correlate such tests with tests of the kind that use the softer material. In addition to that, you can obtain information that is of direct application to the problems of people who desire to cut those harder materials.

O. E. HARDER:[5] In regard to the electromotive force and its constancy, I suggest that tempering the tools for a longer period of time may be of some assistance. That will tend to complete the changes which would take place at those temperatures in the machining operations. Furthermore, within limits in which you can still get a good cutting tool, I suggest that the highest possible tempering temperature be used. If he is going to have temperature developed for cutting tools running to 1500 degrees Fahr., then it is obvious that it is impossible to use tempering temperatures as high as that and still get a good high speed steel. So within a range I think that something may be accomplished by using longer and higher tempering temperatures.

Authors' Closure

The authors are grateful indeed for the manifestation of the spirit in which this paper has been received. We wish to thank those who have contributed to the discussion.

The practice as referred to by Mr. Davis seems to agree in general with the results obtained in the paper. It would seem, however, that a somewhat larger side-cutting angle could be used to advantage.

The comments of Mr. Riegel relating to the analysis of the test log are correct. The bar was purchased as an S.A.E. 2345 steel. The analysis as subsequently determined, however, showed a carbon content of 0.36 per cent, so that the material should have been called an S.A.E. 2335 steel. The phosphorus and sulphur contents should be 0.032 and 0.036, respectively.

The tip of the tool was carefully placed in the plane of the horizontal axis of the work. Because of the short overhang of the bit and the rigid tool holder and mounting, we assumed no deflection. There naturally would be some deflection, but it is believed that working on a diameter of 8 to 12 inches would make a very little difference in the set-up angles being investigated. The cuts taken were relatively light considering that a modern 30-inch swing heavy-duty lathe was used.

In order to keep this paper from being too extensive, no information relative to the power consumption required to cut with the several types of tools or the different steels is presented. This information has been recorded and it is intended to be the subject of a later paper. This work correlates the three components of the cutting force with the temperature developed at the cutting edge as determined by a tool-work thermocouple throughout the cutting life of each tool.

[5]Assistant director, Battelle Memorial Institute, Columbus, Ohio.

Mr. Digges spreads a well-needed word of caution relative to the interpretation of tool-life data when shallow cuts are being taken. It was for this reason that the depth of cut in all cases was determined by measuring the diameter of the log before and after the cut. We were able to keep the depth of cut in our set-up to within 0.001 or 0.002 inch from the intended 0.100 inch. The fact that we were able to duplicate tests caused us to feel that the results obtained are reliable. It was only when new tool bits were introduced that wide variations were obtained. From preliminary tests we were able to select a group of tool bits which seemed to give consistent results and which were used in making the subsequent tests. The fact that this curve, as shown in Fig. 3, shows a smaller value for the $\frac{3}{64}$-inch radius than the $\frac{1}{32}$-inch radius naturally made us feel that one or the other was slightly out of order. The exponent for the 0-inch nose radius tool, however, is well above all others.

Mr. Digges points out that the variation in nose radius of the tools, as shown in Fig. 4, gives a constant of about 100 for the 0-inch nose radius and about 240 for the ¼-inch nose radius tool. It, therefore, appears that the superior performance of the ¼-inch nose radius tool is due to the higher constant rather than the lower exponent. It may be pointed out, however, that the lines representing the cutting speed tool life of the two tools, as given in Fig. 2, are not parallel because of the difference in exponent, so that the performance of the ¼-inch nose radius tool over the 0-inch tool will vary for different values of tool life. In other words, the difference in cutting speeds as represented in Fig. 4 for the 0-inch nose radius tool differs from the others because of the difference in exponent and constant, inasmuch as the exponent in the cutting-speed tool-life equations for the other tools are the same, 111. Their difference in performance is due entirely to the different constant.

The relation pointed out in equation (6) between V, T, and R is confined to the test conditions as reported. We do not at this time know whether this same relation would hold for other test conditions.

Mr. Digges questions the values of cutting speed for a tool life of 300 minutes. We have, on numerous instances, run tool life tests of this duration and found them to agree substantially with the formula resulting from shorter tool life tests. We also have found that in very short tool life tests, the data usually agree with those for longer life, although such data are usually more erratic. It does seem that the failure of tools would be by pure abrasion at comparatively low temperatures for the long tool lives and by high temperature abrasion for the shorter tool life. We have wondered why there was not some more abrupt transition in the line connecting the data of these different zones. It may be pointed out that the size of the cuts in the present paper would not be known as shallow cuts as were those Mr. Digges referred to. They were, on the contrary, midway between the tests of Mr. Digges' shallow cuts and those of roughing cuts of a previous paper.

While this paper was not intended to include tests of variable depth of cut and feed, it is believed from available experimental data that the value of depth and feed does materially influence the value of n in the equation representing cutting speed and tool life. It was not the intention to compare the results obtained in this paper with those obtained under heavier cuts, but rather

to point out the belief that the size of cut has a very distinct influence on the results.

In answer to Dr Styri's question as to why the tests were not run on nose radii greater than ¼ of an inch, it was found that, when using radii larger than ¼ of an inch, chatter was obtained. The chatter was so great that the machine vibrated and it was not possible to get reliable results. That was true also when side cutting angles larger than 30 degrees were used. The curves would indicate that if chatter could be eliminated better results could be obtained. That was not possible under the conditions of these tests.

When the Brinell hardness of the test log is raised to 400 as suggested by Mr. Sadtler, the cost of the log would be much higher, and for that reason we have used annealed test logs of more commonly used materials. We appreciate the suggestion that a harder material should be cut at much reduced speeds resulting in less material being cut up.

The comments of Mr. Sellers are very much appreciated. Undoubtedly, differences between the results of Taylor and those given in the paper are due to the fact that Taylor's tools were of the curved cutting edge type as well as having a large nose radius and his sizes of cut were greater than those covered in this investigation. Unquestionably, further experiments should be carried out even with those tools used in these experiments, with cuts larger and smaller than those reported on. It is expected that heavier cuts will give results which vary somewhat from those reported, while lighter cuts may yield still different results.

The experiments reported on have been conducted with practically no financial support outside the University of Michigan. It seems that tests of this nature are worthy of the consideration of some support from the industries. The authors feel that there is a great deal of work that should be done along this line before complete and final information is available as to the use of single-point tools of high speed steel, Stellite, or cemented carbide.

It is particularly gratifying to note that Professor Wallichs has found that the exponent of the cutting-speed tool-life equation may differ from the Taylor value of ⅛. We do not believe it has been found by American experimentors that the exponent varies with the generally-used steels. The author has shown previously that the analysis of the steel of the more generally used types does not influence the exponents of the feed and drill diameter in the formulas for torque and thrust in drilling, but rather that only the constant is affected as the material is changed. When special steels or other metals, such as cast iron, aluminum, malleable cast iron, etc., are used, separate equations are obtained for each material.

It is of interest, however, to note that Professor Wallich's work does show a variation in the exponent when cuts lighter than those taken by Taylor and the Bureau of Standards are used. This is a subject of an investigation now being carried on.

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